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METALS AND ALLOYS

1

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METALS AND ALLOYS

Volume 7 Number 1 January, 1936

The Magazine of Metallurgical Engineering

PRODUCTION FABRICATION TREATMENT APPLICATION

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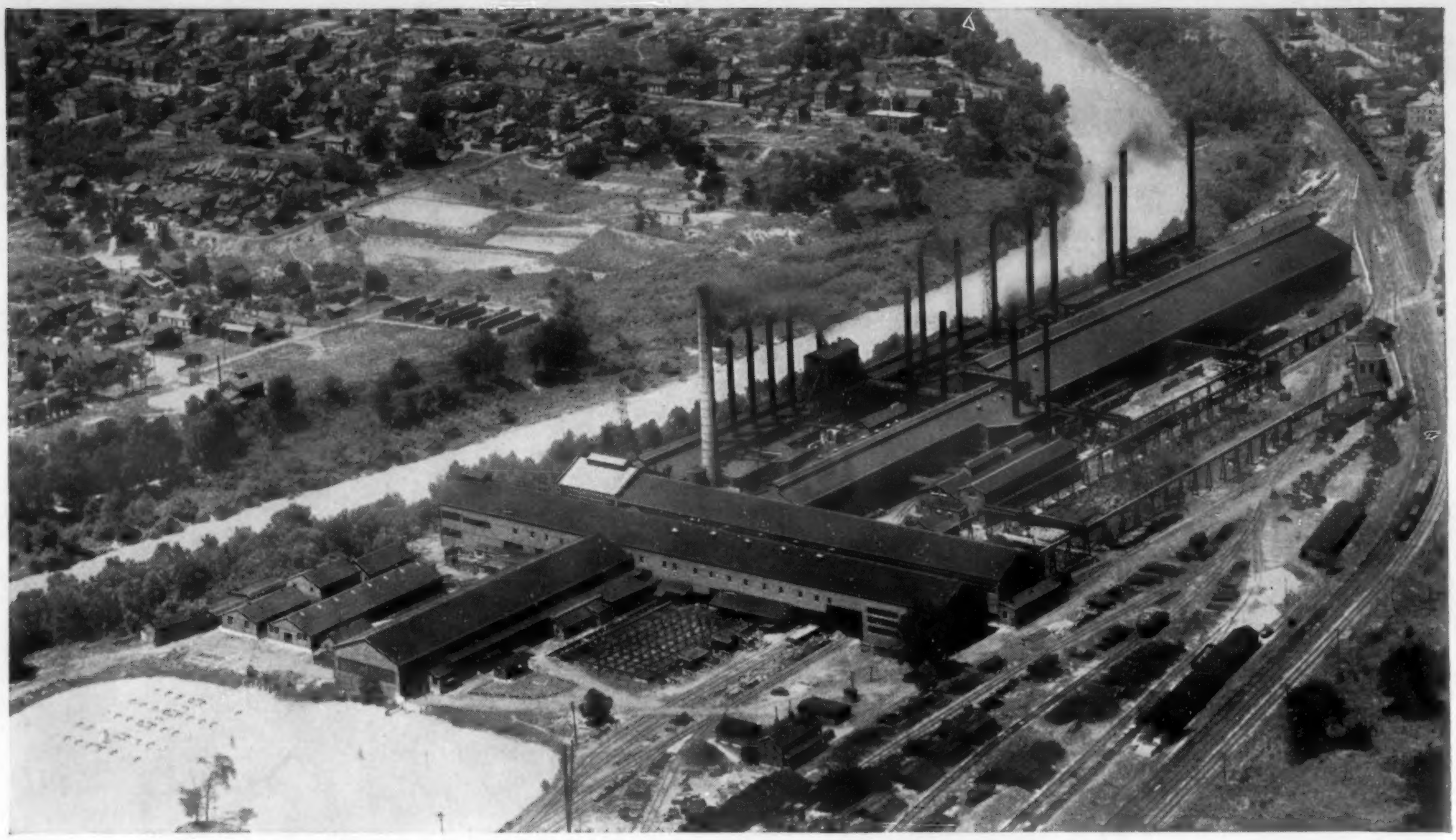
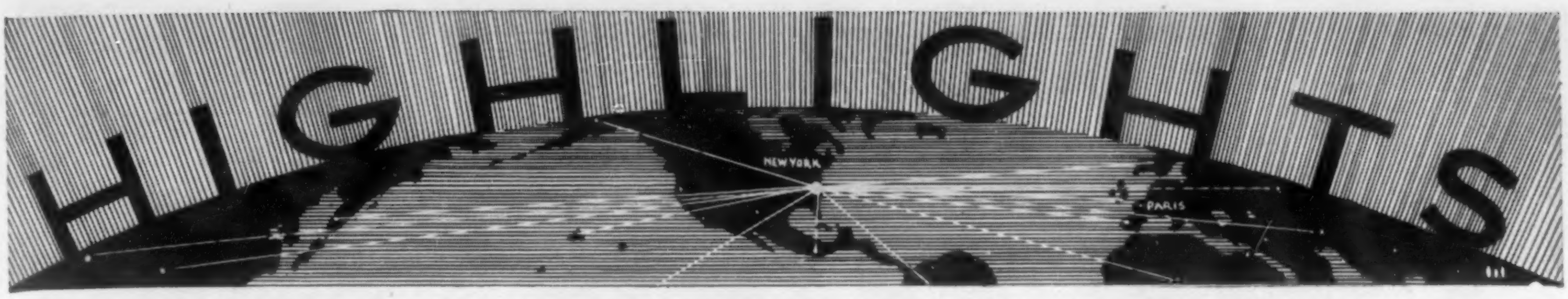
ARTICLES

CURRENT METALLURGICAL ABSTRACTS





Written by the Abstract Section Editors and the Editorial Staff

DO YOU want to know what metallurgical engineers are saying, the world over? Look in the Current Metallurgical Abstracts. Here are some of the points covered by authors whose articles are abstracted in this issue.

Still Discussing Flotation Theory

Flotation theory hasn't "jelled" yet, as evinced by the controversy between Ostwald (page MA 1 L 6) and Wark (page MA 1 L 6), two high authorities. Seidler (page MA 1 L 7) refers to their papers in a discussion.—J.A.

All About Non-Metallic Elements

For a wealth of information on oxides, gases and ingotism, read the "Sixth Report of the Committee on Heterogeneity of Steel Ingots" (page MA 4 L 2)—C.H.H.

Molds and Ingots

All sorts and sizes of molds are discussed by Hruska (page MA 4 R 1). Yakhkind and Gotgelf (page MA 4 L 8) give some interesting points on defects in steel made in round molds, and Honda and Hirone (page MA 4 R 4) talk about thermal stresses in cylindrical ingots.—C.H.H.

Steel Castings

Alloy steel castings and their manufacture are discussed in detail by Hatfield (page MA 4 L 10).—C.H.H.

Heat Treatment of Chromium Steels

Beuloz (page MA 10 R 2) gives heat treatments and properties of a wide range of chromium steels, some of them with such special elements as Mo, W and Ti.—O.E.H.

Normalizing vs. Box Annealing for Sheets

Mort has considered modern furnace design in relation to box annealing (page MA 10 R 7) and normalizing of sheets in relation to cost and properties (page MA 10 R 8).—O.E.H.

Another Step in Heat Resistance

An important advance in the metallurgy of heat resisting alloys is a new alloy for heating elements and other high temperature applications (page MA 13 L 2).—M.H.M.

Wider Use of High Frequency Induction Heating Possible

There are great possibilities for the wider use of high frequency inductive heating on a commercial scale, of, as, and when the first cost of the outfit can be brought down. One sentence in Noble's article, "the cost of a rotating machine was prohibitive," appears in many other evaluations of potential applications of high frequency current. Noble's solution of his own problem, a vacuum tube outfit giving 20 kw. (page MA 13 L 3), is of value as far as it goes. If the output of tubes could be brought up to 100 kw. and over, at reasonable cost of equipment and reasonable upkeep cost, high frequency melting might go places in the plant in competition with other heating methods even on cheap raw materials, instead of merely on the very expensive ones or under very special conditions.—H.W.G.

More About Mg Alloys

Magnesium alloys come in for attention this month, Guzzoni (page MA 29 L 6 & 9) discussing Mg-Al-Ni which he says is mighty poor from the corrosion point of view. Portevin & Bastien (page MA 29 L 8) try to evaluate some Mg alloys as to corrosion in acid and salt media. What designers are figuring on using Mg against such media in service? Why not give the poor things a chance and do our quantitative evaluation of corrosion resistance under such conditions that the results will mean something to the designer?—H.W.G.

Variations in Thermocouple Wire

Roeser and collaborators point out that base metal thermocouple wire varies enough to require calibration of each lot, for anything like precise work, and, in another article (page MA 33 L 2 & 3) tell how to do it.—H.W.G.

Steels in Architecture

Morris and Kautz describe the advantages of stainless steels and enamelled irons for building exteriors (page MA 37 R 6). The magnitude of the market for these materials justifies extensive development and sales work.
—M.G.

Longest Out-Door Welded Steam Line

A 12 in. steam line, 8400 ft. long, 3/8 in. wall thickness, was completely oxy-acetylene welded and erected in the fall of 1934 in about 6 weeks. This line ties in the power plants of the U. S. Industrial Alcohol Co. and the U. S. Industrial Chemical Co., Curtis Bay, Baltimore, Md.

More than 2000 ft. of the pipe were welded and erected on steel structures 30 to 60 ft. in the air. The line traverses filled marshes, and crosses state highways, city streets and a railroad freight yard. A total of 496—12 in. welds, 16—6 in. tee or nozzle welds and numerous smaller welds in the line were made.

The line is used for process steam at 150 lb./in.^{2} and 100° F. superheat. In a test made under full steam pressure no leaks or other failures developed. (page MA 16 L 3).—E.V.D.

Metallurgical Patents

Few metallurgists have the opportunity, time or patience to keep abreast with metallurgical patents. Only the most outstanding cases ever come to the attention of the large body of metallurgical workers. It would seem that a serious attempt on the part of some one to summarize the various metallurgical trends, as reflected in the patent literature, should serve a very useful purpose. An illustration of this is the analysis of German patents showing the development of plating technique in 1934 (page MA 21 L 6).—H.S.R.

Mass Production of Bearings

Any account of Ford's practice for turning out a high-quality product in mass production is sure to attract attention. The production of the steel strip coated with the copper-lead bearing mixture which forms the basis for many of the bearings in the current model car is no exception (page MA 22 L 7). While the research metallurgist may not find anything new in it, the production manager or any one whose job it is to make practical applications will find interesting and valuable suggestions in this account.
—H.S.R.

An Amazing Endurance Ratio

Gough reports the amazing endurance ratio of 0.87 on cast iron (page MA 25 R 2).—H.W.G.

Another Hope Dashed

Wever and Möller (page MA 25 R 4) throw cold water on the idea that broadening of bands in X-ray diffraction tests is an indicator of approaching one-hoss behavior of material under repeated stress. We get so used to finding out that short cut methods for fatigue and creep don't work that we're not even disappointed when such hopes are dashed, and we're still less surprised.—H.W.G.

Two Well-Known Compounds Stigmatized

Intermetallic compounds Fe_3W_2 and Fe_3Mo_2 have been in good standing for some time. However, after re-examination of old X-ray data, Arnfelt and Westgren (page MA 26 L 2) concluded that the compounds are really Fe_7W_6 and Fe_7Mo_6. This will hardly make high-speed steel any worse, but it might contribute to the understanding of the stuff.—J.S.M.

The Editorial Advisory Board

(This Series was Started in the July Issue)

F. N. SPELLER is director of the department of metallurgy and research, National Tube Co., Pittsburgh. A native of Canada, he received the degree of Bachelor of Applied Science and of D.Sc. from the University of Toronto. He was chemist of the city of Toronto from 1894 to 1896, after which he became associated with the Carnegie Steel Co. This was followed by mining engineering work in Alaska and association with the Bureau of Mines of Ontario. He began his long record with the National Tube Co. at McKeesport, Pa., in 1901. In 1904 he became metallurgical engineer for the company, serving in that capacity until 1926 when he assumed his present position.

Dr. Speller's name is synonymous with outstanding research work on corrosion problems and in recognition of this work he has been awarded the Longstreth Medal of the Franklin Institute (1927) and the American Iron and Steel Institute Medal (1931). In 1934 he delivered the annual Howe Memorial Lecture of the A.I.M. and M.E., and has presented many papers on corrosion and metallurgical subjects. His book on "Corrosion" is outstanding.

F. N. SPELLER

As a member of the A.S.T.M. since 1904, Dr. Speller has the unique distinction of having served on Committee A-1 on Steel longer than any other member, and has been very active. He has served on Committee A-5 on Corrosion of Iron and Steel since it was organized in 1905, having been vice-chairman since 1926. His work on other committees—executive (1927-1929), Sub-committee XXI on Welding of A-1, Committee D-1 on Preservation Coating, and D-19 on Analysis of Industrial Waters—has been, and is, invaluable.

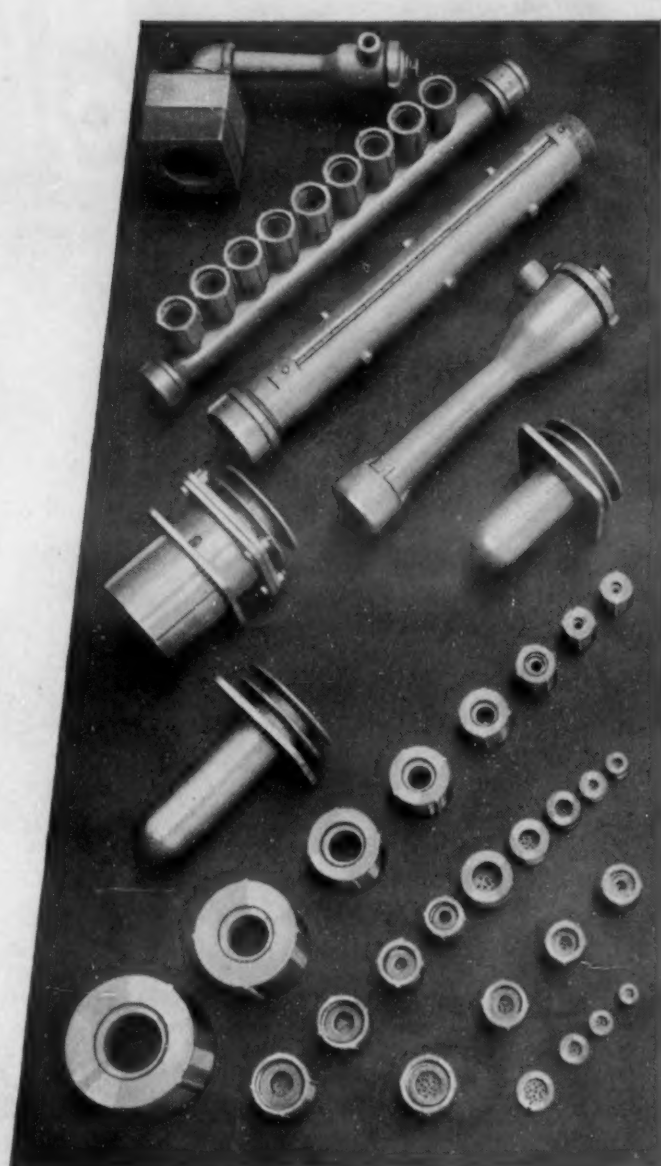
For the past seven years Dr. Speller has been chairman of the Metallurgical Advisory Board that sponsored the eminent cooperative research work in Pittsburgh on the chemistry of steel making.

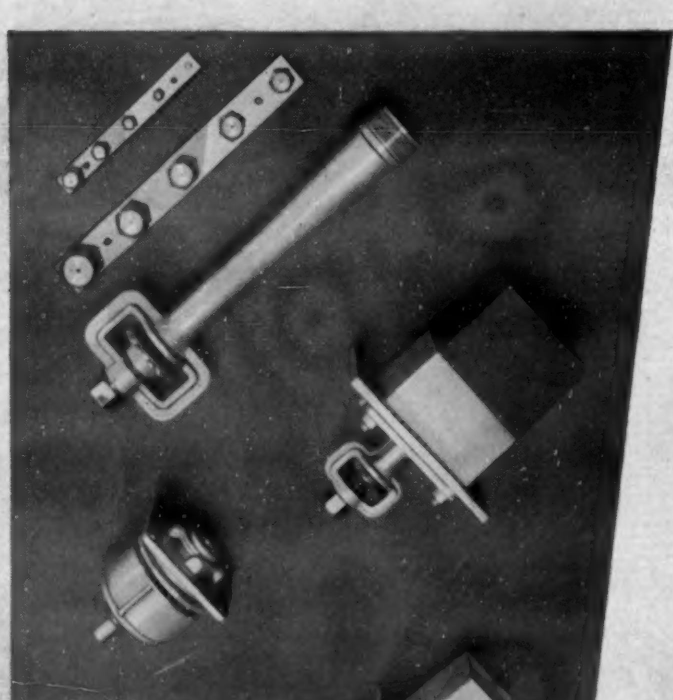
JEROME STRAUSS was recently elected vice-president in charge of research and development of the Vanadium Corp. of America and has been affiliated with that company for the past seven years. He is located at Bridgeville, Pa.

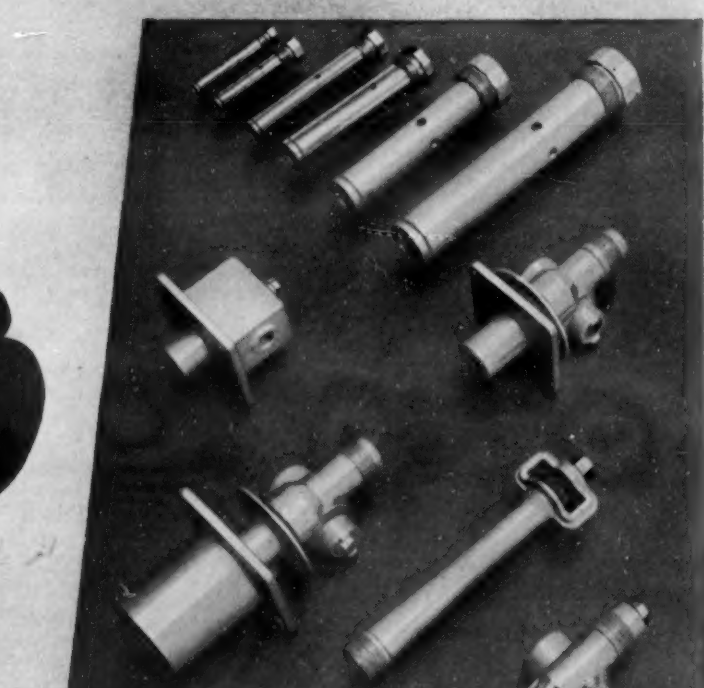
Mr. Strauss was graduated from the Stevens Institute of Technology with the degree of M.E. Following his graduation he went with the Illinois Steel Co., at its South Chicago works, where he remained from 1913 to 1914. From then until 1917 he was identified with the Western Drop Forge Co. as chemist and metallurgist. During the World War, from 1917 to 1919, he was commissioned in the U. S. Army, serving the ordnance department in various capacities. In 1919 he was named chief chemist and later materials engineer for Washington. He went with the Vanadium organization as chief research engineer in 1928.

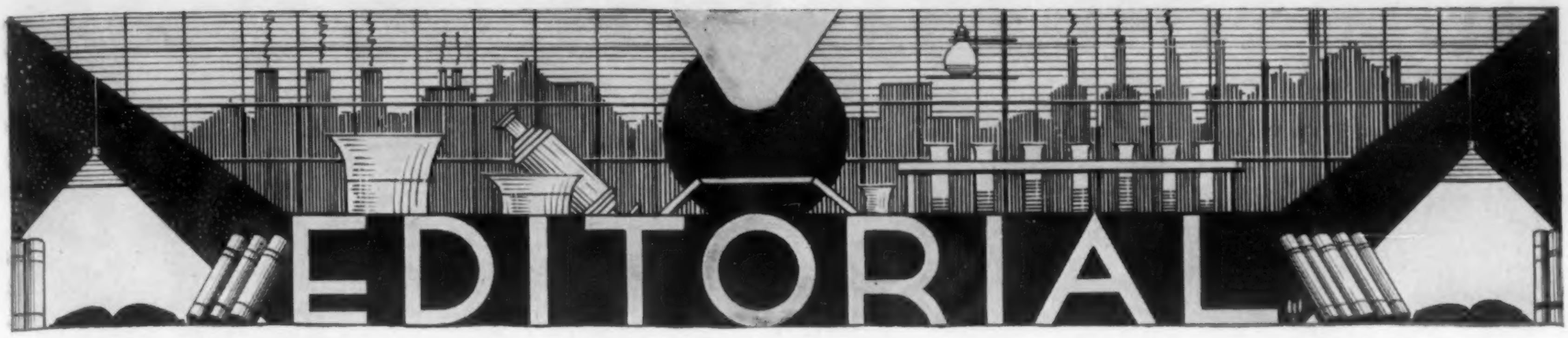
JEROME STRAUSS

Mr. Strauss's numerous society affiliations include the vice-chairmanship of the Pittsburgh chapter of the A.S.M., chairman of the non-ferrous division of the A.F.A., chairman of Committee A-10 on Iron-Chromium, and Iron-Chromium-Nickel Alloys of the A.S.T.M., and membership on the iron alloys research committee of the Engineering Foundation. He holds a commission of lieutenant commander in the United States Naval Reserve.









1935 in Retrospect

WE have never been able to get very excited about annual reviews of progress for publication in the January issue because it takes time to show what has been real progress of consequence from the permanent point of view, and what is just interesting. Hence the real advances of a year are more truly the realization within that time of the importance of developments that started long before, and are still in progress.

Recognition of the fact that, with sufficient further attention, a need is going to be satisfactorily filled and that whatever best fills the need is going to find a market, is at least as important a type of progress and certainly a more clear-cut type, than the springing of some new material or process that has yet to prove its position.

From the point of view of general recognition, certain developments which showed enough of an upswing in '35 to indicate that they are here to stay, might be mentioned.

For use in high-yield strength steels, the cheap alloying elements, especially copper, have come into prominence. It is clear that there is a wide market for such steels, and practically every steel company either has one on the market or is feverishly developing one. There may be room for several sub-species of these steels, but they can't all be best. Sooner or later a degree of automatic standardization will occur in this field. Meanwhile we pity the analysts who have to make routine determinations of so many elements and the superintendents who have to work up the scrap from the present galaxy of high strength steels when it begins to come back. A rather analogous situation exists in the high strength wrought copper alloy field. Everybody has one that's a little bit different from the rest. These are needed, but we probably don't need so many.

With the recent advances in understanding of the principles of grain size control, the old mystery of "body" in steel is evaporating. There is a market for steel of controlled grain size, though the many pages of recent discussion of the subject are rather out of proportion to its commercial importance. There is still something to be done in getting more certain control. An analagous matter, the understanding of the principles of control of graphite size and distribution in cast iron, seems also on the verge of clarification. At least the old idea, that the vital thing was the retention or avoidance of undissolved graphite particles in the melt, is pretty well exploded, clearing the way for more nearly correct theories. The practice of graphite control has advanced and it is clear that there is a market for several grades of cast iron not hitherto made in much quantity. This whole field should show further advance when we know more of the "why" as well as the "how."

The old problem of "gases in metals" is still with us, and some of the discussions of the subject showed little improvement in understanding of the problem, they sounded very much like those of ten years ago. However, some progress seems to be being made in that much attention is focused, both in steel and nonferrous alloys, on the idea that hydrogen, from traces of water vapor, is probably one of the worst actors.

In the field of heat treatment, it is plain that controlled atmospheres are here to stay, and equally plain that lots of the furnaces that are supposed to put out work that needs no pickling aren't doing so all the time. Intensive work on development of better atmospheres and on more uniform operation with the present ones is in order.

The evidence of the last year indicates pretty clearly that the expense and messiness of the old pickle room are not going to be tolerated much longer by the steel industry. Electrolytic pickling bids fair to make pickling much more of a precision operation and one to be carried out as such, under much closer technical supervision. Considerable further development of processes and for wider utilization of those already worked out, will surely come.

The light-weight, porous refractories, with their potentialities for saving of fuel or current, especially in batch furnaces for intermittent operation, have had a very definite degree of adoption. Improvements in their refractoriness and spall-resistance are called for and will doubtless come. Considerable advances have likewise been made in other types of refractories, as commercial acceptance of these shows.

The sources of heat for furnaces now include the radiant tube for gas heating, with some of the advantages of electric heat, though it as yet fits only into a relatively low temperature range. The electric heating industry is countering with resistor alloys suitable for higher temperatures or giving longer life at the usual temperature. Both these advances seem here to stay, with the radiant tube needing alloys suitable for higher temperature source and the resistor alloys needing improvement, either in strength at the higher temperatures at which they are otherwise useful, or better toughness and formability at room temperature, according to their type.

Radiographic inspection of heat-resisting alloys and of pressure vessels and the like has gained greatly in commercial application. The polarizing microscope gives a new means of recognition of the composition of inclusions, a considerable step in their elimination. Electro-diffraction methods are being developed to tell us more about metal surfaces and surface films on metals. Centrifugal casting, especially of heat-resistant alloys, and of cast iron pipe has made notable strides.

The Ford cast crank shaft type of malleable iron, in which copper again enters as an alloying element

(Continued on page 8)

Recent Developments Affecting the Testing Society's Specifications for Steel Castings—I

By R. A. BULL

Consultant on Steel Castings, 541 Diversey Parkway, Chicago

THE author of this series of articles on specifications for steel castings writes with authority and is one of the best-posted men in this field. In substantiation of this, Major Bull was chairman of the special A.S.T.M. committee that functioned under Sub-Committee VIII of Committee A-1 (the society's main committee on steel) which developed the so-called consolidated specifications (A 154), which were made obsolete the past year. He is also chairman of the special A.S.T.M. committee which again is functioning under Sub-Committee VIII of A-1, which developed specifications A 87-35 T and A 180-35 T.

Major Bull has been a member of Committee A-1 and of its Sub-Committee VIII for about 20 yrs. He has also been a member of Sub-Committee XXII of Committee A-1 ever since it was organized. He is also a member of Sub-Committee X on Castings of Committee A-10, which has jurisdiction over iron-chromium, iron-chromium-nickel, and related alloys. Major Bull has also been a member of the A.F.A. committee on specifications for steel castings ever since it was organized. He has also had a broad experience in the production of steel castings.

In all these committees Major Bull has been untiring and devoted in his activity. A clarification of the present situation in this important field is therefore assured.—EDITOR.

THOSE ENGINEERS who would like to know whether significant metallurgical progress has been made recently in steel casting manufacture can get indirectly, but none the less reliably, considerable information on the subject merely by examining the ten specifications for steel castings which were sponsored tentatively by the American Society for Testing Materials during the summer of 1935. Such a scrutiny, when feasible to make, will be an eye-opener for many metal users. Because seven of these specifications were tentatively adopted as recently as August, 1935, (supplementing three specifications tentatively adopted in June of the same year), the average consumer cannot now be assumed to be familiar with them. A considerable period of time may elapse before there is available to some purchasers of steel castings the opportunity of examining the latest purchasing requirements in a volume of the kind annually published by the Testing Society, containing all of its current tentative standards. Another large book that contains each existing set of standard specifications is published by the A.S.T.M. every three years. After the opportunity is afforded for examining the prevailing standard and tentative specifications for a great many engineering materials, considerable time will be required to go through them for the purpose of scrutinizing the major details in all such specifications as cover steel castings. Meanwhile a useful purpose may be served, not only by explaining the newly adopted requirements, but by discussing some interesting metallurgical phases of the subject which have prompted constructive discussions without precedent in Testing Society committees.

Preparing to Pour Arc Furnace Electric Steel in the Foundry. (Courtesy of Michigan Steel Casting Co.)

Exceptional Progress in Developing New Specifications

Probably it will surprise many individuals to learn that ten new A.S.T.M. specifications for steel castings were tentatively adopted during 1935; that 12 such specifications were originated since 1932; and there are now in effect 15 sets of A.S.T.M. specifications covering steel castings. The recent unparalleled activity in establishing acceptance standards for the product of the steel foundry has been brought about by noticeable progress in foundry practice and by the growing dependence of designers on castings.

The preparation of many new purchasing requirements has been accompanied by the serious collective consideration of several problems, including those developed by the design of the casting. There has been plenty of evidence that those who have had experience in making or using steel castings now face, and wish frankly to discuss together, some conditions which arise in the foundry and are entirely beyond the control of the melter. For the first time in the Testing Society's history, significant recognition of some of these factors has been given, through established differentiations in required properties, to be explained later.

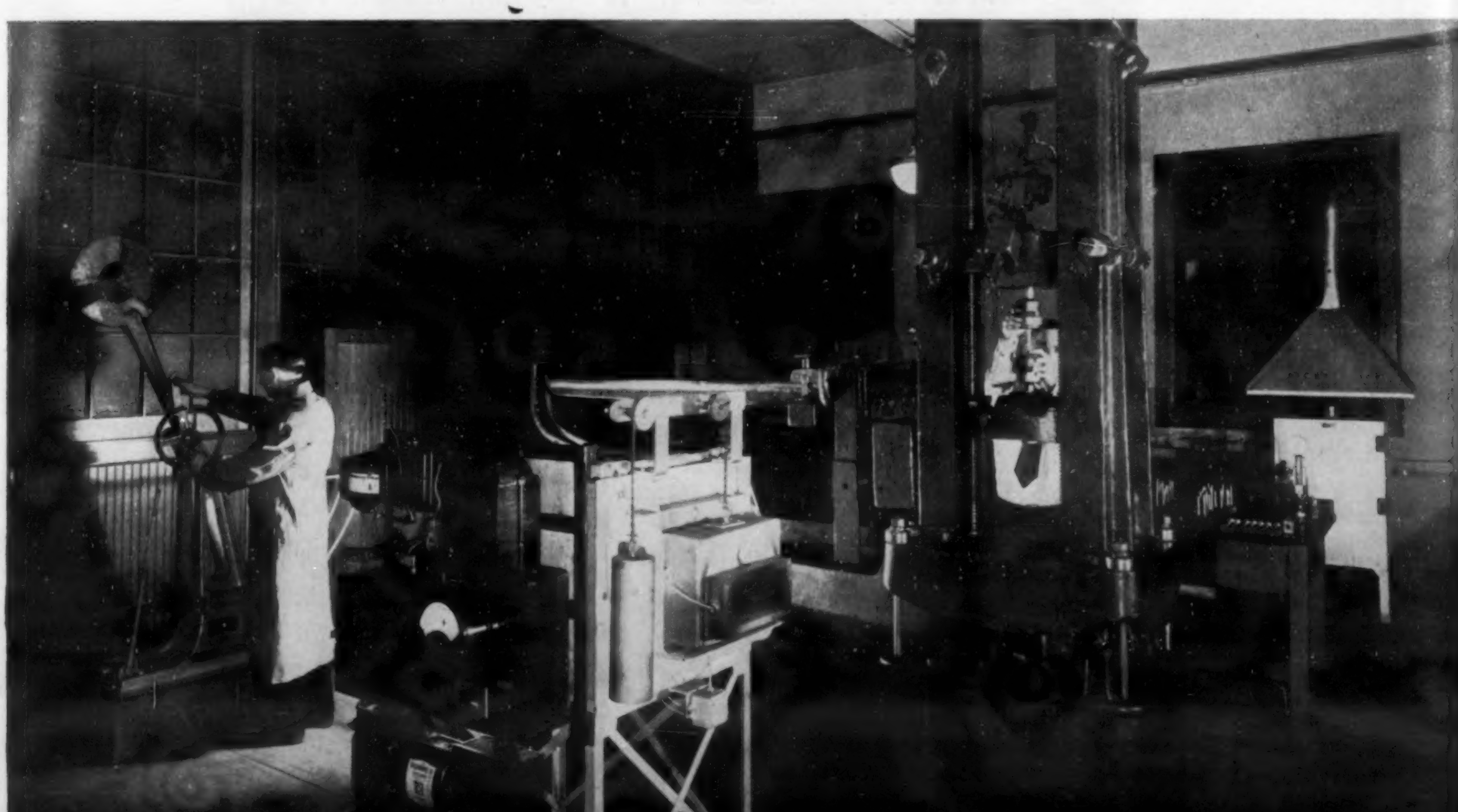
For reasons which the reader may ultimately realize, the author first will describe some formal actions recently taken, before discussing in detail some metallurgical problems whose significance has been demonstrated effectively in commercial manufacture. It is hoped that these features of this contribution will cause readers to regard it as appropriate for publication in METALS & ALLOYS.

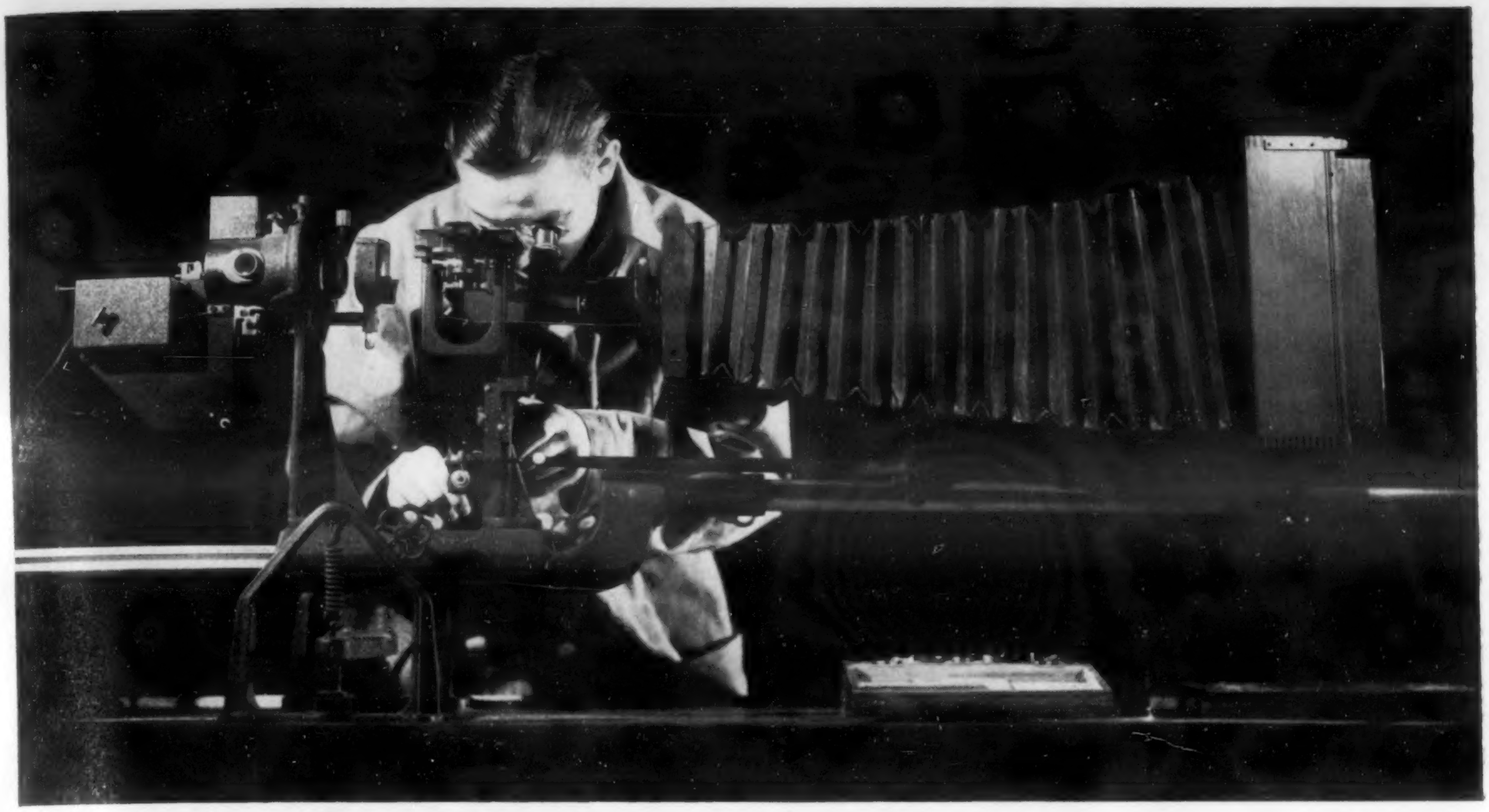
Not until about six years ago did the Testing Society adopt, either tentatively or for use as standard, any purchasing requirements for alloy steel castings. (With justification it can be claimed that this delay was caused more by the average foundryman's former disinclination to express interest in authoritative specifications than by belated progress in manufacture.) In 1930 agreement was reached on tentative specifications which, after some changes, took the form of Standard Specifications A128-33, for Austenitic Manganese Steel Castings. Three years later the society adopted tentative specifications covering eight grades of alloy-steel castings for structural purposes. And during the next year there were tentatively adopted specifications for eight grades of alloy-steel castings for valve and fittings for high temperature service. A significant beginning of the activity that reached a high point in 1935 had thus been made.

Identification Method Used by Testing Society

Those who are not conversant with A.S.T.M. specification designations may be aided in understanding them by a brief explanation of the Society's identification system. It is the practice of this organization, differing from some others, to apply the plural designation "specifications" to each set of purchasing requirements. (This confuses some persons who are accustomed to singularize and to refer to a "specification" as meaning the description of all acceptance standards for one material.) The letter "A" preceding the numerals in the title of specifications indicates that they cover a steel product. The first set of numerals following the letter "A" represent the assigned consecutive number of the class, which is retained for the affected specifications whether or not they are subsequently revised or given either a tentative or standard status. The consecutive number is followed by a hyphen, which in turn is followed by the last two numerals that indicate the year of adoption, or of latest revision if any changes have been made. Specifications having a numeral designation that terminates with the letter "T" have been adopted only as tentative.

Corner in a Steel Foundry Laboratory. (Courtesy of Sivyer Steel Casting Co.)





Examining the Results of Steel Foundry Practices. (Courtesy of Sivyer Steel Casting Co.)

When they are advanced to standard, this letter is deleted.

Cancellation and Replacement of Out-Moded Requirements

The recent adoption of A.S.T.M. specifications for steel castings has been accompanied by the cancellation of three sets of specifications for the product of the steel foundry. One such set, designated as A77-18, was developed in an emergency during the World War, to cover Electric Cast Steel Stud-Link Anchor Chain. These specifications never had a wide application and became superfluous soon after the conflict terminated. The clauses differed from those in other A.S.T.M. specifications for steel castings in many respects, such as the entire absence of chemical requirements. These specifications were discontinued in 1932.

Action of particular interest to founders and users of steel castings was taken early in 1935 for the cancellation of Tentative Specifications, A154-33T. These were originated with the idea that all purchasing requirements for carbon steel castings for miscellaneous, railroad, and marine uses could advantageously be included in one set of specifications. The consolidation plan grew out of the need for uniformity in physical and chemical requirements for identical grades of metal. After two years' trial use, it was decided that the consolidation scheme should be abandoned and that several sets of specifications covering carbon steel castings for service at normal temperatures should be formulated, collectively to serve the purposes of Specifications A154.

While the latter were being developed in 1932 and 1933, substantial progress was made in securing agreement within several organizations on identical or similar standards of acceptance reflecting modern foundry practice. This resulted from bringing into sub-committee conferences the representatives of several national bodies interested in the manufacture or use of cast steel. By the time decision was reached to discontinue Specifications A154, a distinct advantage had been gained through inter-society cooperation. As a result, no time was lost in arranging for assistance from representatives of organizations interested from consuming and producing standpoints, for preparing two sets of specifications that were tentatively adopted in 1935. One set is known as Tentative Specifications A180-35T, for Carbon Steel Castings for Miscellaneous Industrial Uses. The other set is designated as Tentative Specifications A87-35T, for Carbon Steel and Alloy Steel Castings for Railroads. These are intended to take the place, to considerable but not to complete extent, of the so-called consolidated specifications that were tentatively adopted in 1933 and made void within two years' time.

Following the official discontinuance of Specifications A154, and coincidentally with the tentative adoption of the new specifications for general industrial and railroad applications, the cancellation of Standard Specifications A87-27 covering carbon steel castings for railroads was effected. The latter specifications had become technically obsolete. To permit ready reference to modernized requirements by means of a long-established specifications number, the old designation A87 was assigned to the new railroad casting specifications, (the terminals "35T" being added necessarily to make identification complete).

Active cooperation that had been developed between the affected sub-committee of the Testing Society's committee on steel and the committee on specifications for materials of the American Association of Railroads was a factor in causing the A.A.R. specification for steel castings, as issued in 1935 (designated M-201-34) to agree as to all major requirements with the newly adopted A.S.T.M. specifications for the same material. The separate actions of these organizations, taken in harmony with each other, demonstrated the feasibility of securing mutually helpful cooperation between national technical bodies.

One set of A.S.T.M. specifications for carbon steel castings, which in the opinion of many persons has largely outlived its usefulness, still remains "on the

books" of the Society; due to the belief held in some very reputable quarters that such retention for the time being is advisable, pending anticipated agreement on rejection limits for specimens so produced as to be markedly affected by extremely slow cooling. These out-moded specifications are designated A27-24, cover miscellaneous carbon steel castings, and include physical requirements that have not been revised since they were tentatively adopted in 1924.

Carbon Steel and Alloy Steel for Valves and Fittings

Turning now briefly to the so-called steel valve casting specifications, issued by the Testing Society in 1935, a few changes that had no special significance were made in the specifications that were tentatively adopted originally in 1934, and cover eight grades of metal for Alloy Steel Castings for Valves, Flanges, Fittings for Service at Temperatures from 750 to 1100 deg. F. The tentative specifications in their present revised form are designated A157-35T. As the result of experience in trying out the originally specified requirements, the tensile strength and yield point values have been lowered and the ductility values have been raised for the carbon-molybdenum grade, compared with the requirements originally prescribed. Minor changes that were also made in 1935 need not here be enumerated.

The specifications mentioned supplement Standard Specifications A95-33, which cover one grade of metal, for Carbon Steel Castings for Valves, Flanges, and Fittings for High Temperature Service. (It would be helpful to some consumers who are not familiar with Specifications A157 and A95 if the title of the latter were qualified to correspond with an explanatory note now appended to them, stating that this set of purchasing requirements "contemplates temperatures from 500 to 850 deg. F.") No change has been made in these specifications since they were made standard in 1933, following the tentative adoption of the first set of requirements for this product, in 1925.

Alloy Steel for Structural Applications

Tentative Specifications A148-33T cover Alloy Steel Castings for Structural Purposes. These were agreed on in 1933 after several years of effort by a special committee. As previously indicated, they were the first specifications sponsored by the A.S.T.M. for alloy steel castings, with the exception of Specifications A128 covering austenitic manganese steel, originally issued as tentative in 1930. Specifications A148 (structural alloy steel) contain no restrictions for chemical properties excepting maximum limits for phosphorus and sulphur. The three classes that collectively include eight grades of material are established solely by means of the tensile requirements and the described permissible heat treatments. For example, the three grades of one class required to show the highest combinations of physical properties (in one case calling for 150,000 lb. minimum tensile strength per square inch) may be liquid quenched at the option of the manufacturer. Only the grades in one class may be so quenched.

High Alloy Steel Castings

Following many requests from consumers and producers for acceptance standards covering stainless and other varieties of high alloy steels, a sub-committee of Committee A-10 made significant progress during 1935 by developing and obtaining in August of that year the tentative adoption of eight sets of specifications; three sets covering material containing chromium, and five sets covering material containing chromium and nickel. Naturally several of these specifications include requirements totally unlike those applicable to carbon and low alloy steel. Methods for determining corrosion resistance have been outlined for some of the compositions.

In addition to the purchasing requirements for eight grades of high alloy steel castings, specifications were developed in 1935 for 18 and 8 castings by the standing committee that has jurisdiction over cast and wrought ferrous material containing chromium and/or nickel. The result of the letter-ballot taken within this committee prompted decision to defer tentative adoption of 18 and 8 casting specifications, pending further consideration. Probably they will be put in satisfactory shape for action during 1936.

Problems Being Attacked for Modernizing Specifications

With the exception of old Specifications A27 (previously mentioned as being outmoded), all purchasing requirements now "on the books" of the Testing Society are indicative of modern steel foundry practice. This comment should not be construed to mean that all requirements in Specifications A27 are now deemed unsatisfactory for test material *produced under certain conditions among several permissible under these specifications.*

The next installment of this article will deal particularly with the problem just referred to and with another important one that has been given careful committee consideration since the spring of 1935. To be specific, the first of these problems that necessitates solution for the desired amplification of the recently issued Specifications A180 (miscellaneous carbon steel castings) was caused by the fact now generally recognized by experienced producers and consumers that the physical properties in a cast steel specimen are materially affected by the rate at which the test coupon is permitted to cool, not only during heat treatment but while the casting is in the mold on the pouring floor.

The other problem concerns the relationship between chemical composition and weldability. These very practical questions have been attacked intelligently by the carefully chosen group of men, delegated early in 1935 to prepare modern specifications covering steel castings for miscellaneous, railroad, and marine specifications. The latter are to be covered by special specifications that have not yet been developed.

Summarized Specification Requirements

Information of detailed character regarding all such A.S.T.M. specifications as were adopted before July 1, 1935, and regarding additional specifications that were then being voted on for possible tentative adoption may be found in the Cast Metals Handbook, recently published by the American Foundrymen's Association. This volume contains a large amount of recently assembled, accurate data on the products of modern foundry practice.

(To be continued)

NOTCHED BAR TESTING—I

By S. L. HOYT

Director of Metallurgical Research, A. O. Smith Corp., Milwaukee

WHILE the title of this article sounds as if it were meant primarily for the testing engineer, we believe it is of even greater importance to the design engineer. Our reasons for this belief will be brought out in editorial comment when the second installment is published.—H. W. G.

POSSIBLY I SHOULD HAVE CALLED it "Impact Testing" for that is the commonly used name for the type of testing under consideration here. Instead, the present title was selected because it seems worth while to show that the Charpy and Izod tests are not tests which primarily bring out the ability of a metal to resist shocks or impacts but, rather, that *they are tests which show how the metal responds to the stress system produced by a notch.* I don't mean to imply that metals always show the same behavior in static and impact loading—frequently they don't—but rather that the effect of impact belongs in a separate category. I hope that this position will not be construed as being arbitrarily taken, for it seems to me to be dictated by the behavior of metals in static and dynamic loading, on the one hand, and in the plain and notched conditions, on the other. I shall try to make this clear as the impact effect and the notch effect are being discussed.

Impact Tests

When you mention "impact testing" to me, I picture a test which relates to the ability of a metal to withstand impacts, as distinguished from statically applied loads. This statement is in line with a rather general impression that a metal is more apt to fail when it is stressed by impact; that it may be brittle in impact when it otherwise would be ductile; or that when called upon suddenly to absorb a blow, it breaks off sharply instead of deforming. The problem might be stated as follows:-"Is the 'taffy candy' effect reproduced in metal?" The rifle bullet which drills a smooth hole through glass offers no difficulties though perhaps to some it would not be so easy to reconcile their general conception of the effect of impacts with the ability of a steel plate to show ductile behavior to that same bullet. Another way of stating our problem is: Do two steels ever behave quite differently under impact loading even though their tensile properties are the same? The specialist may, and undoubtedly does, have answers to such questions but certainly common knowledge of "velocity sensitivity" is dominated by ideas that are little better than general impressions.

Before the behavior of metals in impact loading can be intelligently tackled from the metallurgical viewpoint, it is necessary to hark back to some simple theoretical mechanics. There we are taught that a "live" load produces twice the stress that the same load would if applied gradually. This live load produces an inertia effect and before it is stopped the stress actually rises to that value. A moving weight (impact) may produce many times the stress of the static load and on that basis it is easy to see that impacts are dangerous, out of proportion to the weight of the moving body. Furthermore the impact is dangerous because, for it to be absorbed, work must be done by producing elastic and possibly plastic deformations in the body. While the stress in static loading depends only on the cross sectional area, mechanics teaches that the stress generated by an impact depends on the length of the absorbing member. In other words, if but a small volume of metal is capable of deforming to absorb energy, the stress must be correspondingly high and the impact is that much more dangerous. It is on this account that adding on metal to a part may, by altering the geometry of the part, make it weaker in impact though it remains equally as strong to static loads.

It is also illuminating to compare the behavior of different materials in impact. While static strength may be compared on the basis of tensile strength, a different yard stick has to be used to compare dynamic strength, for stress is only one component of work. This is brought out nicely by a class room experiment which shows cast iron three times as strong statically as wood but able to absorb only one-third as much energy in impact. When we get to the discussion of the notched bar, we shall see that the reduction in section makes the notched bar break with far less energy absorption than the plain bar, as the latter can deform uniformly over a greater length. These principles are discussed in texts on the strength of materials and the only object in bringing them up here is to indicate what type of impact behavior is covered by mechanics and hence is not our particular concern. In other words there are very good reasons for metals behaving differently in impact than they do in static loading but if the metallurgist is to become concerned, some peculiarity of metallic behavior must be apparent.

There have been various investigations of the relative behavior of metals in static and impact loading, and I am still discussing plain bars of uniform cross section. These tests have been usually restricted to the lower ranges of striking velocity and there it has been found quite generally that the work done to rupture a bar by impact actually exceeds the work to fracture the same material in a static test. The increase is about of the order of 25 per cent but may be

as much as 50 per cent. This increase is held to be due to the necessity for overcoming the internal friction of deformation, which frictional energy is utilized in heating up the part. In the static test, it will be recalled, the load on the test bar is essentially in equilibrium with the resisting forces of the material.

The work done to rupture a tensile bar can be obtained from the area under its stress-strain diagram. If this is done for a series of carbon steels, it is found that the range of composition which gives the maximum work of rupture comes at 0.30 to 0.40 per cent C. This maximum is checked by impact tests, whether single blow or repeated blow, which shows again that the impact behavior may confirm the results of the tensile test. This correlation extends with at least fair accuracy to the reduction of area in the tensile test which is sometimes used as a measure of "toughness." The question of mechanics does not enter here for it is assumed that the same volumes of metal deform in both types of loading. The tests show that it does not make much difference what rate of loading is used, the behavior is very much the same and the work done is also about the same except for internal friction.

We must not assume that all recorded data show the same behavior in static and impact loading. In fact, many years ago, Considere reported that a steel which showed 27 per cent elongation in the tensile test actually shattered on falling on the floor. Later he conducted a series of tensile impact tests on samples of soft iron and found that, above a certain threshold value of the striking velocity, the iron behaved like a brittle metal. The effect of lowering the temperature was also studied and he made the interesting discovery that it had but a negligible primary effect on the energy absorption though it did have the most important secondary effect between 150 and —10 deg. C. of lowering the threshold striking velocity at which brittle behavior set in. I know of no more recent work of this type which might be used to verify this finding.

Sir Robert Hadfield is greatly interested in the effects of high striking velocities on steel and, with Mr. Main, he published some results which included the effects of quite high velocities. They found that projectile velocities of the order of 600 meters per sec. do not produce brittle behavior in ductile steel that is properly grain refined. On the other hand they reported that overheated steel became brittle at a velocity of 560 meters.[1] Zinc and high phosphorus mild steel are probably also in the same category, while Davidenkow reported that overheated boiler plate became brittle at —90 deg. C. while the same material in the standard condition remained ductile down to about —150 deg. C.[2] These tests showed that neither was velocity sensitive, at normal temperatures at least for the striking velocities used, though they both developed brittleness at low temperatures. Work at the Watertown Arsenal, and reported by Mann[3], shows but little difference with common plain carbon and alloy steels, presumably properly heat treated, though the striking velocities reported were all low. The only alloy steel reported which was sensitive to velocity was the 18 and 8 stainless steel which lost one-half of its energy absorbing power as soon as the striking velocity exceeded 8 ft. per sec.

By way of summarizing the effect of impacts, we may say that metals appear generally to repeat their characteristics of the tensile test. The behavior is governed largely by the geometry of the part and by the mechanics of the stress system which develops. This holds even for brittle metals which, insofar as limited data show, absorb about the same amount of energy in both instances. It also appears that in certain important exceptions to this generalization, metals may be brittle to impact though ductile statically. It is unfortunate for a proper understanding of this subject that more and better data are not available. It is particularly unfortunate that the factors are not better understood which cause normally ductile metals to become brittle in impact.

Coming to the behavior of metals when notched we have again to distinguish between simple mechanics and metallic behavior. The differentiation which I have in mind can be explained about as follows. Granted that the presence of a notch in a bar materially alters the behavior of that bar, regardless of what material it is made, is the behavior always predictable on the basis of the mechanics of the notch effect and the known tensile strength and ductility of the material, or can some new aspect of metallic behavior come into play completely to alter such a prediction? If the behavior were always predictable, there would be no point to breaking notched bars for the purpose of studying metal quality. However, it will be shown in the following that our most important material, steel, is only too apt to behave like a brittle metal when notched, though it be perfectly normal in ductility in the tensile test.

The Notched Bar Test

Of all the tests which are used on metals, none has been the subject of so much controversy, both pro and con and private and public, as the notched bar test. This difference of opinion is due to no lack of data, for the test has been studied for over 30 years by competent engineers and metallurgists and the information gained has been given wide circulation. It is not due to lack of sponsorship by reliable parties nor to any demonstrated lack of utility. Those who are aware of this difference in appraisal must be amazed at the survival of any controversy, whatever may be the viewpoint of the individual, and they must be critically interested in an analysis of its reasons or causes.

This controversy appears to me to arise fundamentally from the peculiarities of notch behavior, unless one is willing to assume that some, either deliberately or unwittingly, ignore demonstrable facts. The apparent mercurial character of the notch effect and a rather hazy conception of the subject matter may be contributing factors. This situation is one which typically calls for a guiding theory. Even in its best form this test is but a technological test and deals with purely empirical behavior, whereas a better form of test would develop and measure the fundamental material constants. For another thing we are not accustomed to ascribe failures to lack of notch toughness. Failures due to fatigue, wear, etc., are common enough but to many it would probably be a little forced to say of a failed part, "The metal was strong enough but it lacked notch toughness."

I have intimated at the end of the preceding section that the salient feature of the notched bar test lies in its ability to show brittle behavior when the tensile test returns the metal ductile. Here the test pictures what may happen in service if an unfortunate set of circumstances arises which simulates the notch effect of the test bar. A source of danger as insidious as this would be must be either definitely demonstrated or equally as positively shown to be nonexistent.

Many recorded cases could be advanced here to

throw light on this point but I am using an experiment which I did personally. First, I secured two samples of mild steel, (A and B), each representative of commercial products which are used by the millions of tons. In the tensile tests these two steels were replicas of each other with a yield point of 35,000 lbs. per sq. in., a tensile strength of 60,000 lbs. per sq. in., elongation of 25 per cent and reduction of area of 60 per cent. These figures reflected similar compositions and rolling treatment while it is perfectly obvious that both are plentifully endowed with normal ductility. For the second series of tests I prepared small bars about 1/8 by 1/4 by 2 in. One of these of each steel was bent 180 deg. flat on itself, to confirm the tensile test for ductility. Third, two more bars were notched at the center of the broad faces with a 90 deg. notch about 1/32 in. deep. Steel A was bent in a vise in the direction of opening up the notch until it had bent through a right angle. Then it was straightened out, bent again in the initial direction, and straightened out the second time. Thus it made four right angle bends and broke at the end of the last one with a fibrous fracture.

Now if the tensile test is capable of furnishing a positive criterion of performance by means of either the elongation or the reduction of area, Steel B must behave in the same way in this simple notched bar test as Steel A did. Actually Steel B broke sharply after a couple of light blows and with no measurable deformation. The fracture was crystalline in appearance. In other words the notch brought about a complete reversal in behavior of these two steels in spite of the similarity in behavior in the tensile test. This test was no fluke and was repeated a number of times to demonstrate the performance to interested parties. In fact one-half of the broken bar can be bent 180 deg. Steel A was a well made open-hearth steel and Steel B was a Bessemer steel of the type used for gas pipe. The metallurgist would say that Steel B is of different "quality" from Steel A, whatever that may mean. If you wish to repeat this demonstration for your own experience, do not stop if the first gas pipe does not show the effect as vividly as the one of my test. Steels vary in their behavior when notched and for my example here I give a marked case of notch sensitivity. Remember though that it still showed full ductility in the tensile and bend tests.

It is not necessary, however, to resort to one's own personal experience with notch behavior. Heyn used the test just described as long ago as 1900 to study the effect of over heating on mild steel. Georges Charpy, the eminent French metallurgist, and many others, reported the results of their investigations of the notched bar test to the International Association for Testing Materials at the 1900, 1906 and 1908 meetings. In particular it was pointed out that certain steels showed abnormal brittleness when notched, though the same steels were normal in ductility when tested by the tensile test. This revelation coming from such high authority made a profound effect on many engineers and metallurgists and created intense interest in the notched bar test. At the New York meeting in 1912 the test again received critical attention.

Results which appeared to be as significant as these were bound to be checked. Perhaps the most prominent early attempt to verify these findings was that of the German Society for Testing Materials which appointed a special committee to study the test and report the findings. This committee conducted a long and careful investigation of the notched bar test and, under the chairmanship of Ehrensberger, reported that there were important cases of steel, known to be "meallurgically" in poor "condition," which were pronounced brittle by the notched bar test but which showed normal ductility in the tensile test[4]. Many very illuminating examples could be cited but the figures will not be reproduced here[5]. It was shown in particular that a cast structure or a coarse forging structure conferred upon steel a condition which was decidely low in notch toughness. At a little later date Stagg in this country showed the same low toughness for steel which had been quenched from too high a temperature though the hardness and tensile tests showed no such drop in quality.

In the German work it was the Charpy test that was found best suited to notched bar testing and the large 30-mm. square bar was selected as giving the most reliable results. This test bar has since been the standard for use in Germany. Other countries which have made extensive use of the notched bar test are France, where the test originated, England, where the service records of material supplied for the late World War demonstrated its practical utility, and Italy, where it has been more recently adopted on a large scale for the purchase of railway materials and similar steel products. In all cases which warrant the use of the test as an acceptance test, standard test bars must pass a stated requirement for energy absorption. The result is used in about the same way as the figure for elongation. In the tensile test a steel which came seriously low in elongation would be rejected even though its strength were known to be adequate. The notched bar test fills this function for toughness. In this way the notched bar test provides an overall check on heat and mechanical treatments and on other quality factors which are otherwise not definitely brought out. These tests are made whether or not the part is subjected to "shock" in service.

We should record here that, historically, acceptance of the notched bar test has not been at all unanimous. Perhaps the chief reason for this is that some experimenters report that the results are apt to be decidedly irregular among themselves and that they give one a feeling that the test lacks meaning and precision. On consecutive heats of steel the results are apt to vary considerably though they are made to the same specification and are otherwise replicas of each other. With such uncertainty and fluctuation it is difficult to correlate the results with other properties or with service records. Opinions of this group which rejects the test vary all the way from a feeling that if stress concentrations are severe in service, it is sufficient to use ductile metal, to the feeling that the results do not relate so much to metal quality as to vagaries of the test itself. Furthermore, tensile tests are always carried out which give design values so that an additional test is superfluous. Many feel that if they need to cover toughness in their tests, the reduction of area gives sufficient evidence on that point. This marked disagreement among engineers and metallurgists creates an extraordinary situation in view of the immense amount of work done on the test and it justifies a rather thorough consideration of the notched bar test.

(To be continued)

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EDITORIAL COMMENT

(Continued from page A21)

whose value has only recently been appreciated, has won a place for uses where low impact properties are not serious. The field of composition and structure intermediate among steel, cast iron and malleable iron has thus been opened up. Present achievements are probably only a foretaste of much more to come.

Perhaps this development will lead to a more definite acceptance in engineering applications of the strong, but not very ductile quick anneal malleable. Material of such properties would appear to deserve wider use than it has achieved and one wonders whether it really hasn't enough ductility to serve in places where the engineer still hesitates to use it.

Another old development that is gaining in acceptance, but still falls far short of as wide application as it deserves, is assembly of steel parts by copper brazing in a reducing gas. The method and apparatus are well worked out and commercially available, and the process is so strikingly economical in large production of parts otherwise assembled by far more laborious methods, that it is a mystery why it has not entered many more plants than it has.

One could go on with many other topics of equal interest which have very definitely moved forward in '35, but in any such list no item can be found that really had its birthday in '35, nor is there one that should not show still further advance in '36 and thereafter. Along about '40 and later the things that actually started to brew in '35 will be adding themselves to the list, for we shall then have some idea what they amount to.—H. W. G.

Their Feet on the Ground

IT IS interesting to watch the percolation of science into metallurgical practice through the medium of technical societies. The very early American Brass Founders' Association *Transactions* contain papers and discussion by brass foundry foremen and technical men—usually chemists, as there was little formal metallurgical training in the universities then—which would sound pretty crude when read today. The early ones of the A.S.S.T. weren't always so hot either, but in both cases the groups came together from the earnest desire of the members to help each other.

As scientific knowledge in these lines grew, the scientists became more articulate and the practical men somewhat less so, fearing that their contributions would not look well in comparison with the "high-brow papers." Hence the present embodiments of these groups, the Institute of Metals Division of the A.I.M.E., and the American Society for Metals, eager as they both are for practical papers, have a real task in getting them. Without decrying the value of the highly scientific and theoretical type of paper, there is always a danger that the sunflowers will crowd out the corn.

The American Foundrymen's Association seems to us to strike a very happy medium. They have kept up such a balance that neither a high-brow article nor one on some simple shop kink is felt to be out of place at their sessions. These men have their feet on the ground.

Another organization that bids fair to come in this category is the American Electroplater's Society. The electroplating industry has been a rule of thumb recipe sort of business and it has taken much missionary work on the part of Blum, Phillips, Hogaboom and others of that ilk, to make the average plater realize that plating can, and should, be a science instead of an art. But the average plater, especially in the larger plants, is eager for technical facts, and likes to know why a thing happens as well as that it does happen.

Actually, science is taking its place in plating at a rather amazing rate. The bulk of the papers at the last convention of the Electroplater's would do credit to any technical society. And they have the added merit of being so expressed that they are informative and understandable both to the high-brow and to the experienced, but not technically-trained, plater. The platers are beginning to talk in technical language, the case being quite similar to the extension of the open-hearth man's understanding and vocabulary, due to Herty's work and to the Open-Hearth Conferences.

That the plating industry is putting its recently disseminated knowledge to good use is evidenced by a remark of Prof. Vuillenmeir of Dickinson College at the last convention. He said that the only example of a peeled nickel coating, made in production, he could show his students was on the back bumper of the car belonging to the Professor of Greek, and, as the car was 10 years old, he was afraid it might be turned in on a new one some time!

The papers at the Convention showed a keen appreciation of the importance of the metallurgy of the metal to be plated, and of problems of adhesion and of diffusion. Too few metallurgists consider that plating is a metallurgical process, they tend to feel that it is in the chemist's domain. We believe that plating will soon be considered as the important metallurgical processes it really is and that a closer hook-up between metallurgists and platers would be all to the good.—H. W. G.

Insulated Open-Hearth Furnaces

IT is now only a few years ago when the suggestion was made to insulate open-hearth steel furnaces. Since then the movement has gained decided favor. The wonder is that a scheme, so apparently sensible and simple, had not been tried earlier. Insulation to conserve heat is not a new idea. It is applied in many ways in industry. In the operation of open-hearth furnaces without insulation, the gradual eating away of linings results in excessive loss of efficient heat, in the intake of surplus air, and in great discomfort in hot climates to workmen. One reason for the probable reluctance to insulate such furnaces was the supposition that insulation would increase the temperatures inside the furnace to such an extent that damage to refractories would be intensified and the life of the furnace shortened.

Pioneer work by one open-hearth operator in particular, whose confidence and enthusiasm spread to others, soon resulted in the refutation of such doubts or fears and in the establishing of the practicability of insulation on a sound basis. A large number of plants are now using insulation in one way or another—on parts of the furnace and checkers or on the entire structure. The benefits and economies, as testified to at the annual meeting of the Open-Hearth Committee of the American Institute of Mining and Metallurgical Engineers, are evidently firmly established. In the opinion of one prominent open-hearth superintendent, within a few years furnace operators who do not insulate their furnaces will be behind the times. —E. F. C.

Phosphorus as an Alloying Element in Low Carbon, Low Alloy Steels

An Experimental Study

by C. H. LORIG and D. E. KRAUSE

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IT IS EVIDENT from the literature, which has recently been reviewed in these columns[1], that, under suitable conditions, and in suitable company, phosphorus is a useful alloying element in steel. It seems especially interesting for possible use in the low cost, low alloy, high yield strength steels that are so much in the eye of the engineering and metallurgical public at present. Indeed, in one of these rather complex steels, "Cor-ten," phosphorus already appears as one of the essential alloying elements, and obviously serves a useful purpose there.

The literature is lacking in information to indicate, with sufficient definiteness, in what other low alloy combinations phosphorus would deserve experiment. There is evidence that:

1. Phosphorus notably raises yield and tensile strengths, as do most alloying elements, and correspondingly lessens static ductility. These effects follow fairly smooth curves up to around 1 per cent P.
2. Somewhere in the neighborhood of 0.3 per cent P the impact resistance, tested at room temperature, drops sharply. This imposes a practical limitation upon the amount of phosphorus that can be utilized in steels for shock-resistant uses. The effects of carbon and alloy content, and of grain size, upon this knee in the impact curve, are not clearly established.
3. In company with copper, phosphorus improves the resistance of a steel to atmospheric corrosion. Its effect by itself and in other alloy combinations, needs clarification.
4. Phosphorus steels may have some interest on account of their electrical and magnetic properties.
5. Since phosphorus raises the critical range, the proper finishing temperatures and heat-treating temperatures will differ from those of steels low in phosphorus, so the critical temperatures need further study.

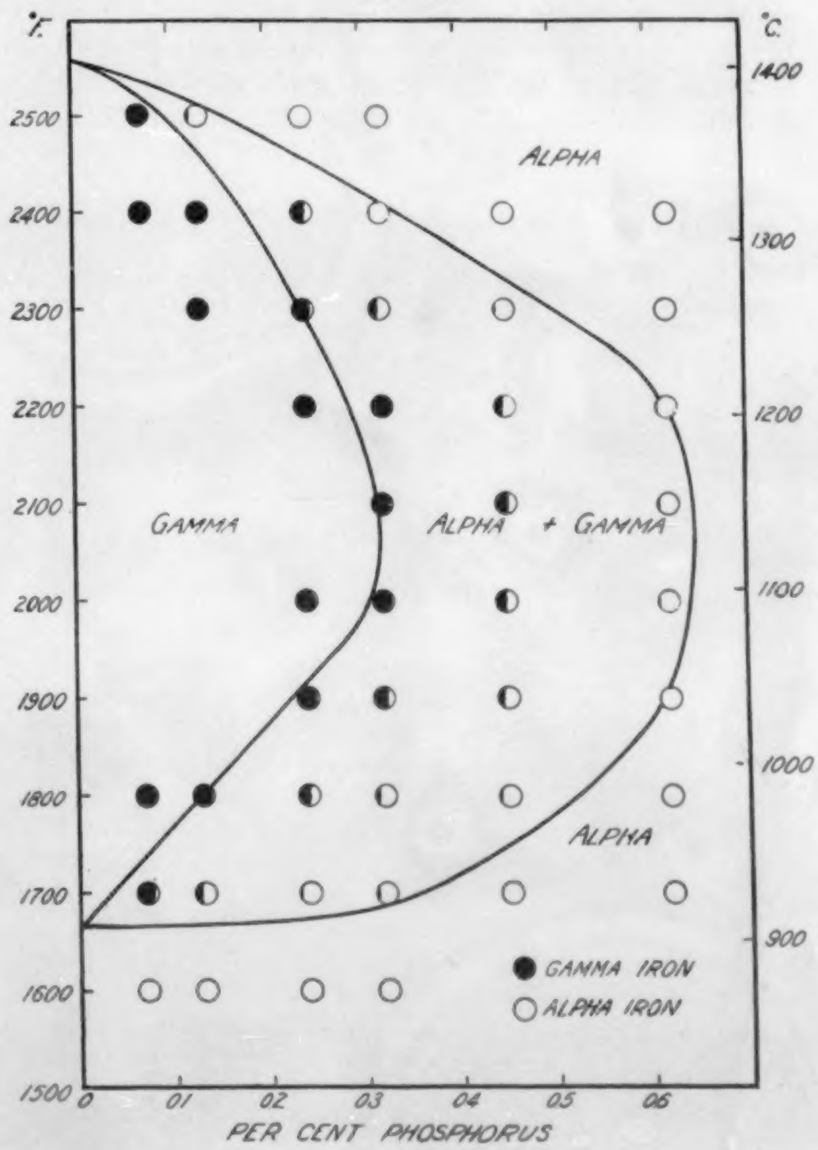
The Monsanto Chemical Co., of Alabama, formerly the Swann Chemical Co., is, as a producer of ferrophosphorus, interested in obtaining information along these lines that might be of assistance to those interested in cheap alloying elements and hence it commissioned Battelle Memorial Institute to obtain some of the needed information. The following is a progress report on some of the findings to date. No attempt was made in these studies to obtain data on phosphorus in cast steels.

In view of the size of the experimental ingots as against production ingots, it is conceivable that rate of cooling in the ingot may so affect phosphorus distribution that the reported results may be only a qualitative guide to large scale production, so that an added factor may need to be taken into account in commercial practice. That such may be the case is suggested by Bennek's[2] recent studies. However, if it is established by further work on a larger scale that the phosphorus limits indicated by the laboratory work are shifted a trifle by rate of freezing, and the auxiliary problem as to proper handling of rate of freezing on a production scale arises, this would not affect the validity of the data given for the rate of freezing that was involved.

The Effect of P on the Gamma Loop

As was brought out in a previous publication[1], the data of Haughton and those of Oberhoffer and Esser, as reported by Vogel, are not in agreement as to the limits of the gamma loop. Haughton reports that the gamma field extends to about 0.50 per cent P around 1950 deg. F. while the data reported by Vogel show

Fig. 1. The Gamma Loop in Steels of 0.925 C, 0.10 Si, 0.15 per cent Mn.



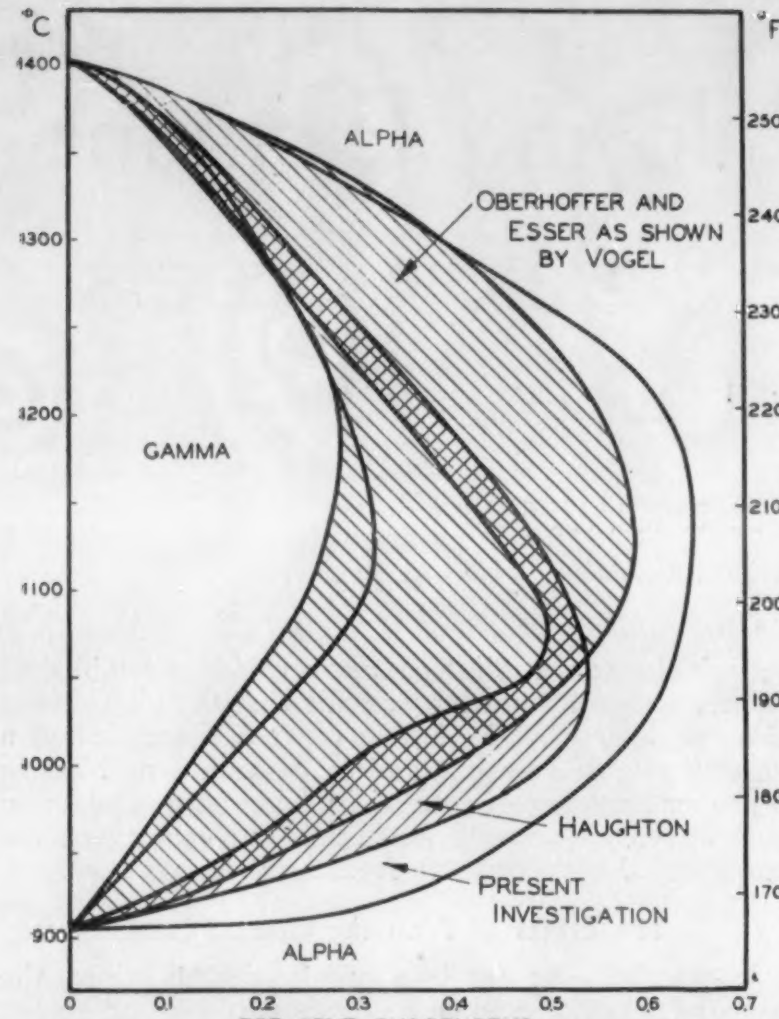


Fig. 2. The Limits of the Gamma and Alpha Plus Gamma Fields for Carbonless or Very Low Carbon Alloys, as Determined by Different Investigators. Haughton's materials contained no manganese.

it to extend only to about 0.30 per cent at 2050 deg. F.

Specimens of varying phosphorus content and with about 0.025 per cent C, 0.10 per cent Si, 0.15 per cent Mn and 0.03 per cent S were heated to various temperatures in a platinum wound furnace for 30 min. and dropped directly from the furnace into water, then examined metallographically. The results are shown in Fig. 1. Fig. 2 shows that these data agree reasonably well with those reported by Vogel, but not with those of Haughton, whose alloys contained no manganese and probably contained slightly less carbon.

To examine the conditions toward the tip of the gamma loop, an alloy of 0.025 per cent C, 0.18 per cent Mn and 0.45 per cent P was water-quenched from 1900 deg. F. (1040 deg. C.). In agreement with Haughton, this showed a duplex structure, Fig. 3. Another alloy of 0.024 per cent C, 0.17 per cent Mn and 0.62 per cent P was quenched from the same temperature. According to Haughton's curve for alloys free from Mn and C, this should be single phase alpha, but the limit was not definitely determined by him, since he studied no alloys between 0.47 and 0.69 per cent P. As Fig. 4 shows, this contained a very small amount of gamma. Grain growth was extreme in this specimen so treated, and some of the larger grains, Fig. 5, show markings along cleavage planes, possibly Neumann lines formed during quenching, or possibly a separation of phosphide though, according to the equilibrium diagram, this would not be present at this composition save in the case of very considerable segregation. Alloys with such a structure are, of course, extremely brittle.

Effect of Carbon Plus Manganese

Larger amounts of carbon and manganese, such as would be met in commercial steels, would be expected to extend the gamma loop. The location of the pure gamma field, (end of Ac_3 range) in steels of around 0.15 (0.12-0.18) per cent carbon, 0.16 per cent silicon and 0.49 to 0.67 per cent manganese, was determined by quenching and metallographic examination as had been done for the lower carbon range, and also, for some of the samples, by dilatometric analysis, the specimens being heated in the dilatometer at 7.5 deg. F. per min. It is seen from Fig. 6 that the gamma field is extended so that, with around 0.20 per cent C and 0.60 per cent manganese, it is possible to have pure gamma with around 0.45 per cent P. The duplex alpha-plus-gamma field extends, in the steels of higher C and Mn contents, considerably beyond the limits

Fig. 3. Structure of an Alloy of 0.025 C, 0.18 Mn, 0.45 per cent P, Water Quenched from 1900 deg. F. Magnification, 100 diameters. Etched with 2 per cent nitric acid in alcohol.

Fig. 4. Structure of an Alloy of 0.024 C, 0.17 Mn, 0.62 per cent P, Water Quenched from 1900 deg. F. Magnification, 100 diameters. Etched with 2 per cent nitric acid in alcohol.

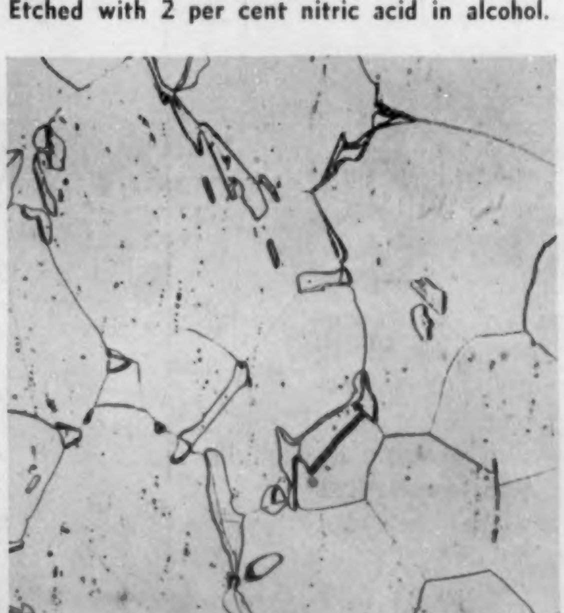
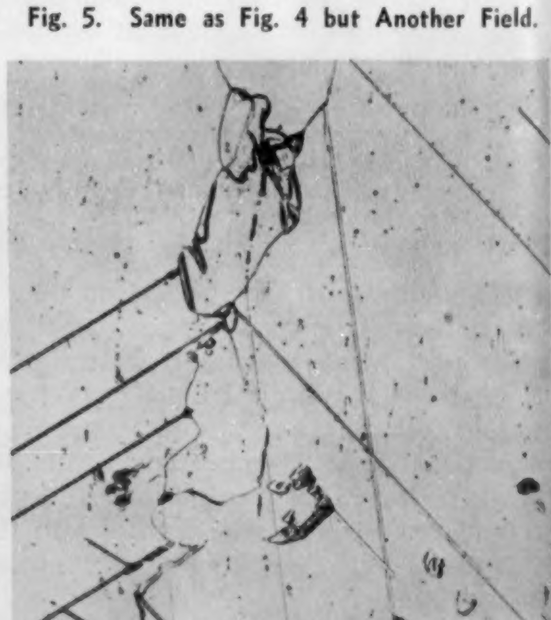


Fig. 5. Same as Fig. 4 but Another Field.



shown in Fig. 1 for steels of low C and Mn. The structures of steels containing respectively 0.07 per cent C, 0.49 per cent Mn and 0.60 per cent P, and 0.12 per cent C, 0.61 per cent Mn and 0.68 per cent P, quenched from 1900 deg. F., are shown in Figs. 7 and 8. The amount of the light constituent, ferrite, shows that in steels with such carbon and manganese contents the duplex range may be expected to extend well beyond 0.70 per cent P.

It will be noted from Fig. 1 that the temperature interval in which gamma is transformed to alpha widens notably as phosphorus increases above the content normally found in commercial steels. The alloys beyond the limit of the alpha-plus-gamma field, i.e., those that have remained pure alpha and no part of which has been through the grain-refining inversion, are definitely impact-brittle. The knee in the impact curve shows up at phosphorus contents somewhat below the limit of the pure gamma field, so impact brittleness cannot be solely ascribed to the retention of some primary alpha, i.e., to the failure of some grains to have ever been gamma. It seems possible, however, that the first alpha to form in alloys whose phosphorus content is close to the limit of the gamma loop, has too long a time for grain growth in its passage through the wide alpha-plus-gamma field on cooling. It is suggested that an examination of the width of the alpha-plus-gamma loop in phosphorus steels, carrying other alloying elements, might be interesting in view of the possibility of a correlation between the extent of this temperature interval and the behavior of such alloy steels in impact. If embrittlement is related to the width of the alpha-plus-gamma interval, then the thing to be sought is such a combination of alloying elements as will restrict this interval.

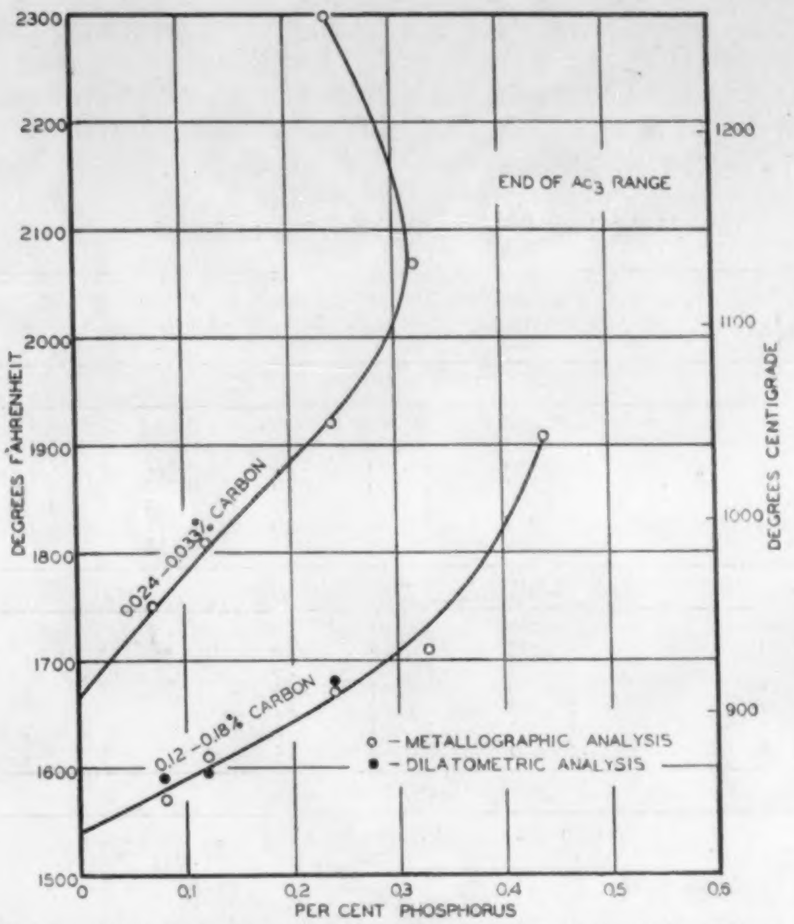


Fig. 6. Shift in Location of the Gamma Loop (End of Ac_3 Range) by Increase in Carbon from About 0.025 to about 0.15 per cent and in Manganese from About 0.15 to About 0.60 per cent.

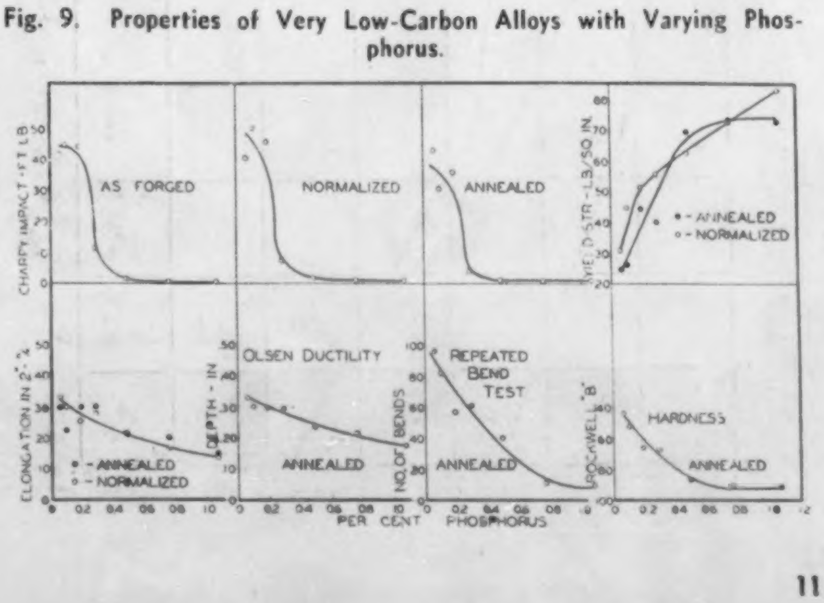
Fig. 7. (Left). Alloy of 0.07 C, 0.49 Mn, 0.60 per cent P, Water Quenched from 1900 deg. F. Magnification, 100 diameters. Etched with 2 per cent nitric acid in alcohol.

Fig. 8. (Right). Alloy of 0.12 C, 0.61 Mn, 0.68 per cent P, Water Quenched from 1900 deg. F. Magnification, 100 diameters. Etched with 2 per cent nitric acid in alcohol.

Static and Impact Properties of Low Carbon, Phosphorus Steels

It was at first thought that the drop in impact resistance would be greatly accentuated by increase in carbon and that chief attention needed to be paid to steels of as low a carbon content as can practically be attained, as in "ingot iron" practice, so a series was made up to show the change in ductility due to phosphorus from 0.06 to over 1 per cent in very low carbon

Fig. 9. Properties of Very Low-Carbon Alloys with Varying Phosphorus.



alloys. The alloys contained 0.01 to 0.02 per cent C, 0.05 to 0.06 per cent Mn, 0.02 to 0.07 per cent Si, about 0.03 per cent S with P varying as shown in Table 1. The steels were melted in a magnesite lined high-frequency induction furnace, using ingot iron melting stock and ferrophosphorus.

TABLE I.—Change in Ductility Due to Phosphorus

P, %	Rockwell B Hardness	Tensile Strength p.s.i.	Yield Strength p.s.i.	Elongation, Per cent in 2 in.	Olsen Ductility, in.	No. of Repeated Bends
			ROLLED, BOX ANNEALED			
0.06	44	42,000	24,800	30	0.330	96
0.10	52	43,900	26,000	22½	0.300	82
0.19	66	56,100	44,500	30	0.297	57
0.29	67	57,700	40,000	30	0.295	61
0.49	86	77,000	69,200	21	0.235	40
0.76	91	83,200	73,100	20	0.214	11
1.07	91	85,500	72,400	15	0.175	10
			ROLLED AND "NORMALIZED"			
0.06		41,600	31,000	33		
0.10		49,600	45,000	30		
0.19		57,200	51,600	26		
0.29		64,100	55,800	28		
0.49		71,600	62,300	22		
0.76		81,300	72,800	17		
1.07		90,800	83,000	14		
			IMPACT—CHARPY—Ft.lb.			
		As Forged	"Normalized"		Annealed	
0.06		42	41		43½	
0.10		45	50½		31	
0.19		44½	46		36½	
0.29		11½	7½		4	
0.49		1½	2		1	
0.76		½	1		1	
1.07		½	1		1	

The small ingots were hot forged at about 2250 deg. F., reheated to 1550 deg. F. and rolled to 22 gage sheet, with four reheatings during rolling. After pickling in 5 per cent H_2SO_4 the sheets were box annealed at 1325 deg. F. for 8 hrs., hydrogen being passed in at the end of this period and during 4 hrs. furnace cooling to 1000 deg. F. The sheets were cooled from 1000 deg. F. in air. The repeated bend tests were made by clamping a strip 1 in. wide between two ½ in. dia. steel rods in a vise and bending the strip by hand 90 deg. to one side of the vertical, then 90 deg. to the other side till fracture occurred.

Sections of the ingots were forged to ½ in. by 3½ in. and Charpy impact test made in the as forged, annealed and "normalized" conditions. For these specimens the annealing was for 10 hrs. at 1325 deg. F., furnace cooled to 900 deg. F., then air, and the "normalizing" in air after 3 hrs. at 1625 deg. F. These "normalizing" temperatures were, as Fig. 1 shows, not high enough for true normalizing of any but the lower phosphorus alloys, so the whole set of data represents material but slightly altered from the as-rolled condition. The finishing temperatures in rolling and forging the higher phosphorus alloys were also too low for proper hot-working according to the data of Fig. 1. In spite of these variations from what are probably more suitable temperatures for handling the higher phosphorus compositions, the static properties follow rather smooth curves, as is shown in Fig. 9, with no sharp break in static ductility, but the dynamic ductility falls sharply at 0.30 per cent P.

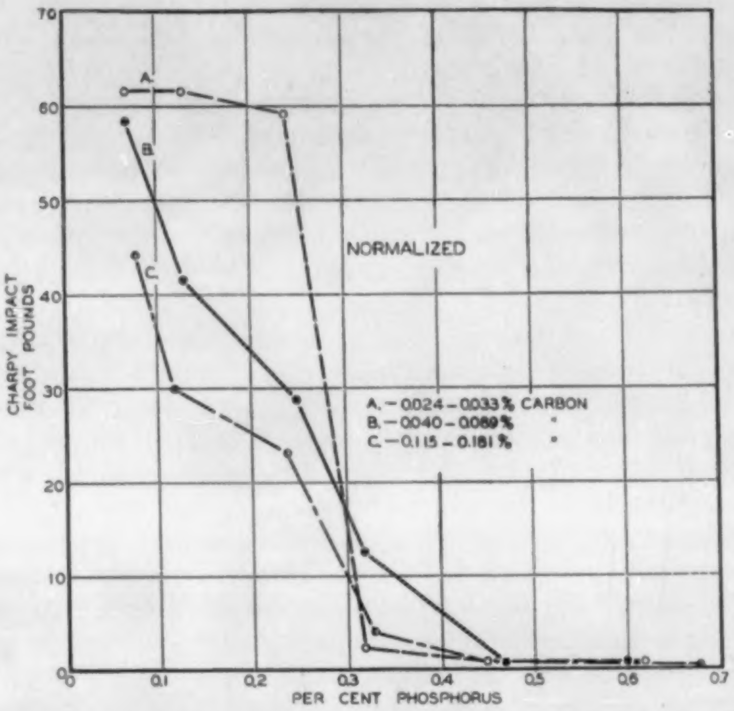


Fig. 11. Impact Properties of Some of the Steels of Table 2—Normalized.

The Effect of Carbon and Aluminum Upon Impact

Other series were made up with higher manganese and higher carbon, as shown in Table 2. Boiler

Fig. 10. Impact Properties of Some of the Steels of Table 2—As-Rolled.

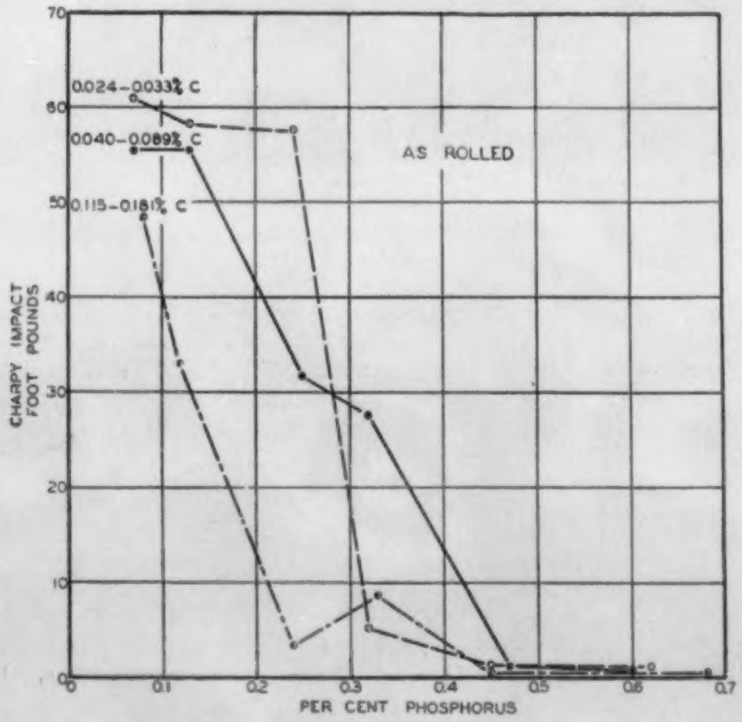
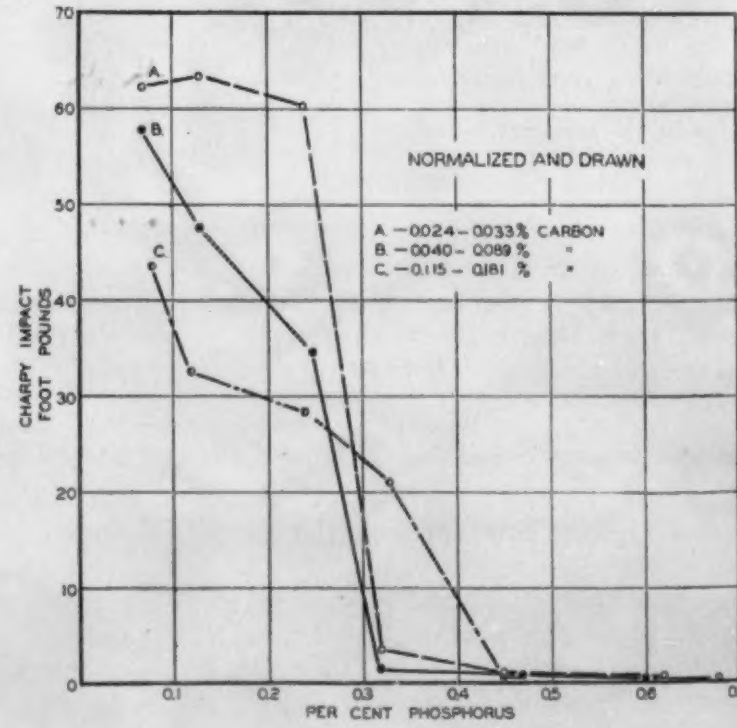


Fig. 12. Impact Properties of Some of the Steels of Table 2—Normalized and Drawn.



punchings as well as ingot iron were used as melting stock to produce the higher carbon steels. These were hot-rolled and tested as-rolled, also normalized, and normalized and drawn as indicated in the Table.

TABLE II.—Charpy Impact Data and Heat Treatment Data on Iron Phosphorus Alloys with 0.02 to 0.18 deg. C.

							Charpy Impact, Ft. Lbs.			
							Normalized		Normalized and Draw	
Spec. No.	C %	P %	Al %	Mn %	Si %	As Rolled	Treat.	Impact	Treat.	Impact
1	0.02	0.07		0.14	0.10	60-62	B	63, 61	B&D	63, 63
2	0.03	0.13		0.15	0.10	58, 59	B	61, 63	B&D	64, 62
3	0.02	0.24		0.17	0.10	59, 55	C	59, 59	C&D	61, 59
4	0.02	0.32		0.18	0.10	5, 5.5	C	2.5, 2.5	C&D	2.5, 5
5	0.02	0.45		0.18	0.10	1.5, 1.5	H	1.0, 1.0	G	1.5, 1.0
6	0.02	0.62		0.17	0.10	1.0, 1.5	H	1.0, 1.0	G	1.0
7	0.04	0.07		0.20	0.10	56, 54	B	57, 59	B&D	59, 56
8	0.07	0.13		0.47	0.10	54, 57	B	39, 44	B&D	46, 49
9	0.07	0.25		0.48	0.10	31, 32	B	29, 29	B&D	38, 32
10	0.04	0.33		0.35	0.10	30, 26.5	C	23, 2.0	C&D	1.5, 2.0
11	0.09	0.47		0.63	0.10	1.5, 1.0	C	1.0, 1.0	C&D	1.0, 1.0
12	0.07	0.60		0.49	0.10	1.0, 1.0	H	1.0, 1.0	G	0.5, 1.0
13	0.15	0.08		0.67	0.16	47, 49	A	41, 47	A&D	44, 43
14	0.18	0.12		0.63	0.16	32, 34	A	30, 29	A&D	32, 33
15	0.14	0.24		0.49	0.16	3, 3	A	22, 23	A&D	28, 29
16	0.16	0.33		0.64	0.16	16, 1.5	B	7, 1.5	B&D	21, 21
17	0.14	0.45		0.63	0.16	1.0, 0.5	C	1.0, 1.0	C&D	1.0, 1.0
18	0.11	0.68		0.61	0.16	1.0, 0.5	H	0.5, 1.0	G	1.0, 0.5
19	0.13	0.23	0.05	0.56	0.12	30, 27	A	32, 30	A&D	35, 34
20	0.14	0.22	0.10	0.64	0.12	32, 32	A	30, 30	A&D	35, 37
21	0.16	0.22	0.15	0.68	0.12	30, 30	A	31, 31	A&D	32, 34
22	0.11	0.23		0.98	0.11	10, 12	A	10, 8	A&D	32, 30
23	0.14	0.23		1.01	0.11	23, 24	A	25, 22	A&D	27, 26
24	0.14	0.23		1.44	0.11	28, 23	A	23, 24	A&D	29, 32

Description of Heat Treatments

(A) Heat ½ hr. at 1700 deg. F., air cool.
(B) Heat ½ hr. at 1800 deg. F., air cool.
(C) Heat ½ hr. at 1900 deg. F., air cool.
(D) Heat 4 hrs. at 1250 deg. F., air cool.
(G) Heat ½ hr. at 1900 deg. F., air cool, heat 4 hrs. at 1250 deg. F., air cool.
(H) Heated 4 hrs. at 1300 deg. F., air cool.

The impact results are shown in Figs. 10, 11, 12. The drop at about 0.30 per cent P for the low carbon steels was again in evidence. Increase in carbon lowers the impact resistance and, in the as-rolled condition, brings the sharp drop down to about 0.25 per cent P. By heat treatment the sharp drop can be shifted back to about 0.30 per cent P.

At the end of the table three steels with 0.23 per cent P and 1 to 1:5 per cent Mn are included, which gave better impact than the corresponding steel with only 0.5 per cent Mn.

The impact resistance of the steels with 0.22 to 0.25 per cent P was very markedly improved by a grain-

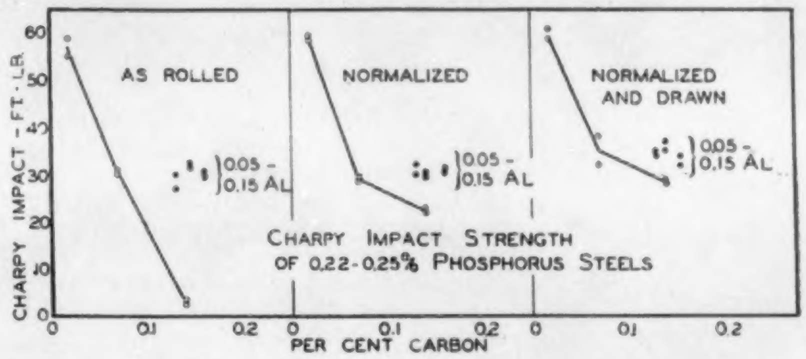


Fig. 13. Effect of Carbon, and of Grain-Refining Additions of Aluminum, upon Impact Resistance of Steels with 0.22 to 0.25 per cent P.

refining addition of aluminum, as Table 2 shows. The effects of carbon and of aluminum are shown in Fig. 13. Aluminum additions are of such importance that in considering the results described below it is necessary to note whether the steels have or have not been aluminum-treated.

(To be continued) p 51-56

References

[1] H. W. Gillett. "Phosphorus as an Alloying Element in Steel." METALS & ALLOYS, Vol. 6, October, 1935, page 280, November, page 307.
[2] H. Bennek. "Einfluss des Phosphors auf die Anlassprödigkeit." *Archiv für das Eisenhüttenwesen*, Vol. 9, Sept. 1935, pages 147-154.

A Note on Stress Relieving Monel Metal

A Contribution from the Watertown Arsenal

AVAILABLE DATA on stress relief annealing Monel metal indicate that 274 to 302 deg. C. (525 to 575 deg. F.) should be used. When checking these temperatures with high strength Monel metal plate, it was found that higher temperatures might better be used.

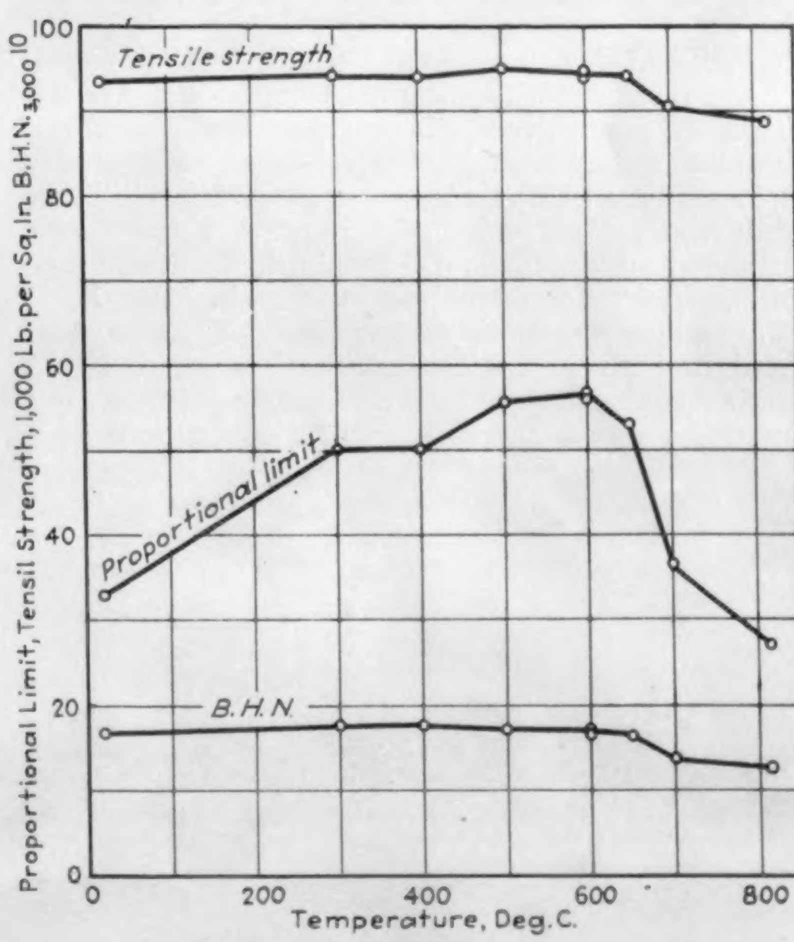
The accompanying diagram shows the effect upon tensile strength, proportional limit and hardness of stress-relief annealing rolled plate for 3 hr. at the indicated temperatures. The analysis of the plate was:

	Percent		Percent
Cu	29.10	C	0.18
Ni	67.29	Al	0.18
Fe	2.08	Si	0.08
Mn	1.09	Mg	0.004

The properties of the plate, as received, were 93,500 lb. per sq. in. tensile strength, 33,000 lb. per sq. in. proportional limit, 37.9 per cent elongation and 65.0 per cent reduction in area, with a Brinell hardness number of 163.

The highest proportional limit of this high strength plate was realized if stress relief annealed near 600 deg. C. (1112 deg. F.), rather than 300 deg. C. (572 deg. F.).

The length of time which the metal is held at temperature affects the results too long, such as 8 hr., lowering the proportional limit markedly.



It is probable that the amount of cold work which the metal receives prior to stress relief annealing will govern the maximum temperature to be used.

A Study of Al-Hg-Zn Anodes in Acid Zinc Plating Baths

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ABSTRACT: A study of Al-Hg-Zn anodes in a $ZnSO_4 — Al_2(SO_4)_3 — NH_4Cl$ bath at Colorimetric pH values of 2.0 to 4.0 and three other acid zinc baths of commercial composition at a pH of 4.0 has been made. Temperatures as high as 50° C. (122° F.) and anode current densities as high as 300 amp./sq. ft. (32.4 amp./sq. dm.) have been employed. Current efficiencies, anode polarization and sludging tendency, bath voltages and pH variations, and character of deposits are reported upon. A comparison with commercially pure zinc anodes is made.

This is a contribution from the Hanson-Van Winkle-Munning Co., Matawan, N. J.

THE Al-Hg-Zn ANODE was just introduced in 1932 as an improved anode for both acid and alkaline zinc plating baths.[2] At that time its performance was reported for a zinc sulphate bath containing aluminum sulphate and ammonium chloride and operated at a colorimetric pH of 4.0, a current density of 20 amp. per sq. ft. (2.2 amp. per sq. dm.) and room temperature. A comparison with mercury zinc and straight zinc (commercially pure) anodes was also made. The Al-Hg-Zn anode was superior with respect to efficiency, freedom from sludging, freedom from chemical attack by the bath, the ability to reduce the pH changes in the bath and the quality of deposits produced.

More recently this anode has been successfully employed in acid baths continuously plating wire at much higher current densities. Freedom from sludge, a considerable saving in metal, otherwise lost, and ductile zinc deposits of excellent protective value have been obtained under these more rapid plating conditions. In view of the fact that rapid continuous plating of wire and strip is receiving increasing consideration, this study of the Al-Hg-Zn anode in acid baths to determine its characteristics under a wide range of plating conditions was undertaken.

Part I. The Aluminum Sulphate Bath at Different pH Values

IN Part I of this study, Bath I-C (Table I) was made up in three quantities, the pH being adjusted to 2.0, 3.0 and 4.0 respectively. Meta cresol purple was used for values under 3.0. Higher values of pH were not employed because the aluminum precipitates voluminously above a pH value of 4.5 and such baths were considered unsatisfactory. Sulphuric acid and ammonium hydroxide were used in adjusting the pH.

Three liter quantities of the three pH baths were placed in battery jars and connected in series with a copper coulometer. Fresh portions of solution were employed for each run and the usual precautions were taken in the preparation and operation of the coulometer.

In determining the properties of the anode, one anode was placed midway between two sheet steel cathodes, at a distance of 3.25 in. (8.25 cm.), the cathodes having the side adjacent to the jar insulated. Current densities from 20 to 100 amp. per sq. ft. (2.2 to 10.8 amp. per sq. dm.) were employed, the total ampere hours for each run averaging about 380.

Sand cast elliptical anodes containing 0.23 per cent Hg, 0.3 to 0.5 per cent Al, and the balance high purity commercial zinc were prepared with copper wires securely fastened in one end. When immersed to a depth of 5 in. (12.7 cm.), the area was 0.1 sq. ft. (0.93 sq. dm.).

The bath voltage and pH were determined at the end of each run. The anode polarization was determined in a separate run, allowing the electrode to reach equilibrium for each current density before making the measurement. The method used for measuring the polarization was the same as that described in previous studies.[3]

The anode efficiency was determined by comparing the loss in weight with the copper deposited in a cou-

TABLE NO. I.—$ZnSO_4$ Bath Compositions***

Bath No.	$ZnSO_4.7H_2O$ g/L	$ZnSO_4.7H_2O$ oz/gal	$ZnSO_4.7H_2O$ N	$Al_2(SO_4)_3.18H_2O$ g/L	$Al_2(SO_4)_3.18H_2O$ oz/gal	$Al_2(SO_4)_3.18H_2O$ N	NH_4Cl g/L	NH_4Cl oz/gal	NH_4Cl N	pH**	Additions g/L	Additions oz/gal	Additions	Operating Temperatures deg. C.	Operating Temperatures deg. F.
I	240	32	1.7	30.	4	0.27	15.	2	0.3	4.	....	....	None	50	122
I-C†	240	32	1.7	30.	4	0.27	15.	2	0.3	2., 3., 4.	....	....	None	32 to 42	90 to 108
II	240	32	1.7	30.	4	0.27	15.	2	0.3	4.	7.5	1.0	Licorice	50	122
III	405	54	2.8	30.	4	0.27	15.	2	0.3	4.	56.3	7.5	Na_2SO_4	50	122
IV	360	48	2.5	..	..	...	30.	4	0.6	4.	15	2	$NaC_2H_3O_2 3H_2O$	50	122
											120	16	Dextrose		

* Baths II, III and IV are essentially those recommended by Blum and Hogaboom, "Principles of Electroplating and Electroforming," 2nd Ed. (1930).
** Colorimetric pH. Metacresol purple for values under 3.0. Brom phenol blue for values 3.0 and higher.
† Bath I-C same as I except for pH and lower temperature of operation.

lometer in series with it. Any film on the anode was brushed off and included in the loss in weight. The *precision* of the measurements was better than ± 0.5 per cent.

The character of deposit and cathode efficiency were determined from 20 to 80 amp. per sq. ft. (2.2 to 8.6 amp. per sq. dm.) using sheet steel cathodes 0.1 sq. ft. (0.93 sq. dm.). In each case the plating time was sufficient to give a deposit approximately 0.00075

TABLE NO. II.—Variation of Anode Efficiency and Polarization with C. D. and pH in $ZnSO_4$ Baths Using Al-Hg-Zn Anodes*

Bath Composition **		I-C		I-C		I-C	
		pH Colorimetric					
Bath I-C		2.0		3.0		4.0	
C.D., amp. per sq. ft.	Temp., Deg. C.	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts
20†	32-42	99.4	0.10	99.3	0.11	99.3	0.03
45	32-40	99.6	0.22	99.6	0.55	99.7	0.11
80	32-42	99.0	0.48	99.1	0.27	99.2	0.19
100	36-42	99.2	0.42	98.9	0.30	99.0	0.07

* Sand-cast zinc anodes containing 0.3 to 0.5% Al and 0.25% Hg.
** Bath compositions in Table No. I.
† 380 amp. hr. runs average.

in. (0.019 mm.) in thickness, assuming 100 per cent cathode efficiency.

The precision of the cathode efficiency measurements was better than ± 0.5 per cent except in the few cases where the loose crystals formed at the edges of the cathodes. The precision in such cases was probably better than ± 1.0 per cent, since the quantity of crystals was small and usually sufficiently adherent to be readily handled. Where crystals fell from the cathodes on to the watch glasses upon which the plates were placed, they were washed, dried and weighed with the cathodes in question.

Discussion of Results

The results discussed below apply to a $ZnSO_4$-$Al_2(SO_4)_3$-NH_4Cl bath (I-C, Table I) which was operated at a colorimetric pH of 2.0, 3.0 and 4.0 and at temperatures between 30 and 42 deg. C. (90 and 108 deg. F.) using Al-Hg-Zn anodes. The variation in temperature was caused by the unavoidable heating of the baths at the current densities employed, but was somewhat less for the lower current density runs than that indicated.

TABLE NO. III.—Variation of Cathode Efficiency and Character of Deposit with Current Density and pH in $ZnSO_4$ Bath Using Al-Hg-Zn Anodes*

Bath Composition **		I-C		I-C		I-C	
		pH Colorimetric					
Bath I-C		2.0		3.0		4.0	
C.D., amp. per sq. ft.	Temp., Deg. C.	Eff., per cent	Deposit †	Eff., per cent	Deposit †	Eff., per cent	Deposit †
20	30-32	96.7	SWF	98.5	SWF	98.8	SWF
30	30-32	98.0	SWF	98.9	SWF	99.7	SWF
45	30-32	99.4	WCE	98.9	WCE	100.0	WCE
80	30-32	96.5 ‡	WCE	99.4	WCE	100.7	WCE

* Sand-cast zinc anodes containing 0.3 to 0.5% Al and 0.25% Hg.
** Bath compositions given in Table No. I.
† All deposits at least 400 ampere minutes—0.0005 in. at 100% efficiency.
‡ Some loss of fine loose crystals.
Legend: SWF = smooth white fine grained.
WCE = white crystalline edges.

The anode efficiency (Baths I-C, Table II) under the above conditions is between 98.9 and 99.7 per cent for current densities between 20 and 100 amp. per sq. ft. (2.2 and 10.8 amp. per sq. dm.). It is also independent of pH between the values of 2.0 and 4.0.

The anode polarization (Baths I-C, Table II) increases slightly as the current is raised. The increase is from 0.1 to 0.42 volts at a pH of 2.0, but not more than from 0.03 to 0.19 volts at a pH of 4.0 for current densities between 20 and 100 amp. per sq. ft. (2.2 and 10.8 amp. per sq. dm.).

The cathode efficiency (Baths I-C, Table III) between 20 and 80 amp. per sq. ft. (2.2 and 8.6 amp. per sq. dm.) in unagitated baths is slightly higher at any current density in the 4.0 pH bath than for the lower ones. At a pH of 4.0, the efficiency increased from 98.8 per cent at the lowest current density to 100.7 per cent at the highest. A similar increase in efficiency with current occurs in the lower pH baths. In general the efficiency is between 96.7 and 100.7 per cent giving an average value that is about equal to the anode efficiency.

The deposits (Baths I-C, Table III) are uniformly smooth, white and fine grained up to 30 amp. per sq. ft. (3.2 amp. per sq. dm.). At this current density there was some slight evidence of pitting, which was not apparent at higher or lower current densities. At 45 amp. per sq. ft. (4.9 amp. per sq. dm.) and above the deposits are white with loose fine crystals at the edges.

The bath voltage (Baths I-C, Table IV) varies little with the pH, but increases with increasing current density as would be expected. For the cell arrangement employed the pH 4.0 bath requires approximately 1 volt at an anode current density of 20 amp. per sq.

TABLE NO. IV.—Variation of Bath Voltage and Acidity with Current Density and pH in $ZnSO_4$ Bath Using Al-Hg-Zn Anodes*

Bath Composition **		I-C		I-C		I-C	
		Colorimetric pH at start †					
Bath I-C		2		3		4	
C.D. Anode, amp. per sq. ft.	Temp., Deg. C.	Bath, volts	pH change	Bath, volts	pH change	Bath, volts	pH change
20	32-42	1.11	0.4	1.07	0.8	1.08	0.1
45	32-40	1.97	0.8	1.94	0.0	2.00	0.1
80	32-42	2.70	0.2	2.66	0.2	2.94	0.0
100	36-42	7.70	0.1	7.75	0.3	8.00	0.0

* Sand-cast anodes containing not over 0.5% Al and 0.25% Hg.
** Bath compositions in Table No. 1. Bath volume = 3 and 4 L in cold and hot baths respectively.
† Meta cresol purple for 2 pH bath and Brom phenol blue for 3 and 4 pH baths. All pH changes are positive unless otherwise noted.

ft. (2.2 amp. per sq. dm.) and as high as 8 volts at 100 amp. per sq. ft. (10.8 amp. per sq. dm.).

The pH changes (Baths I-C, Table IV) during plating are negligible in the pH 4.0 bath. In the pH 2.0 bath the pH increases during an 18-hr. run at low current densities. At higher current densities the change in pH is negligible.

Conclusions

In conclusion it may be stated that a bath of composition I-C gives the best results at a pH of about 4.0. No advantage is to be gained by operating at a lower pH, and at much higher values the tendency for aluminum to precipitate is a serious disadvantage.

The Al-Hg-Zn anode, when used with the bath I-C at a pH of 4.0, at temperatures between 30 and 42 deg. C. (90 to 108 deg. F.) and current densities of 20 to 100 amp. per sq. ft. (2.2 to 10.8 amp. per sq. dm.) has an anode efficiency of 99 per cent, a polarization under 0.2 volt and is sludge-free.

The cathode current density must be below 45 amp. per sq. ft. (4.9 amp. per sq. dm.) if satisfactory, adherent deposits are to be obtained, although agitation of the bath or motion of the work may be expected to improve this somewhat. Where the current density is below the limiting value mentioned, the cathode efficiency will be about the same as the anode efficiency, i.e., 99 per cent.

The bath voltage will be approximately 2.0 volts when the current density does not exceed 45 amp. per sq. ft. (4.9 amp. per sq. dm.), and the change in pH of the bath during plating will be negligible.

Part II. Commercial Acid Zinc Baths at Elevated Temperatures

SINCE it has been shown that Bath I-C is not capable of producing satisfactory deposits at current densities as high as can be used at the anode, the following changes suggested themselves as offering some possible improvement at the cathode, while still investigating the anode performance:

First: To raise the bath temperature, keeping the bath composition the same (Bath I, Table I).

Second: To add licorice in an amount commonly recommended for the aluminum sulphate type of bath (Bath II, Table I).

Third: To investigate other bath compositions, such as the Na_2SO_4 and the $NaC_2H_3O_2$ type of baths (Baths III and IV, Table I.)

All four baths are of commercial compositions.[4] They were prepared in quantity and the pH adjusted to a value of 4.0 colorimetric (3.34 quinhydrone). Four liter portions were then placed in battery jars and connected with a coulometer in the manner described in Part I.

The procedure was similar in every respect to Part I except that the anodes were immersed from 3 to 5 in. (7.6 to 12.7 cm.) to give convenient anode current densities. The temperature was maintained at 50 deg.

TABLE NO. V.—Variation of Anode Efficiency and Polarization with Current Density and Bath Composition in Acid Zinc Baths Using Al-Hg-Zn Anodes*

Bath Composition **		I		II		III		IV	
Baths I-IV		pH Colorimetric 4.0		4.0		4.0		4.0	
Temp., Deg. C.	C.D.,	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts
50 ± 2	100†	99.2	0.09	99.0	0.06	99.2	0.23	99.1	0.14
50 ± 2	150	99.0	0.10	98.9	0.08	99.1	0.23	98.9	0.14
50 ± 2	200	99.2	0.07	98.9	0.10	99.1	0.22	99.1	0.14
52 ± 2	300	98.7		98.0		98.4		99.1	

* Sand-cast zinc anodes containing 0.3 to 0.5% Al and 0.25% Hg.
** Bath compositions given in Table No. 1.
† 260 ampere hour runs average.

C. (122 deg. F.) by placing the jars in a water bath. The anode current density was varied from 100 to 300 amp. per sq. ft. (10.8 to 32.4 amp. per sq. dm.) and the cathode current density from 80 to 150 amp. per sq. ft. (8.6 to 16.2 amp. per sq. dm.).

Discussion of Results

Baths I-C and I: It is interesting to note the effect of raising the temperature by comparing the results obtained with Bath I-C (Tables II to IV) and Bath I (Tables V to VII). The anode efficiency remains practically constant at 99.1 per cent in spite of the higher temperature and current density employed. The anode polarization is less, not exceeding 0.1 volt, and as before, no sludging occurs. The cathode efficiency is lowered, but satisfactory, smooth, white deposits are obtained at 80 and 100 amp. per sq. ft. (8.8 and 10.8 amp. per sq. dm.) without any agitation of the bath.

Baths I to IV: The results obtained with the four high temperature baths, I to IV, (Table I) are fully recorded in Tables V, VI, VII. Rather than attempt a lengthy written comparison, the data have been summarized in Table VIII by averaging the values obtained for all current densities. This table, therefore, represents a rough comparison of the four baths when operated with Al-Hg-Zn anodes at anode current densities between 100 and 300 amp. per sq. ft. (10.8 and 32.4 amp. per sq. dm.) and cathode current densities from 80 to 150 amp. per sq. ft. (8.6 to 18.2 amp. per sq. dm.).

TABLE NO. VI.—Variation of Cathode Efficiency and Character of Deposit with Current Density and Bath Composition in Acid Zinc Baths Using Al-Hg-Zn Anodes*

Bath Composition **		I		II		III		IV	
Baths I-IV		pH Colorimetric 4.0		4.0		4.0		4.0	
Temp., Deg. C.	C.D., amp. per sq. ft.	Eff., per cent	Deposit †	Eff., per cent	Deposit †	Eff., per cent	Deposit †	Eff., per cent	Deposit †
50 ± 2	80	88.7	BSW	93.8	DS	94.3	GS	92.9	WS
50 ± 2	100	83.3	BSW	94.4	DS	89.1	GS	89.9	WS
50 ± 2	150	97.2	BTE	89.1	DTE	94.5	GTE	98.0	WTE
50 ± 2	150‡	100.1	WPT	88.4	DS	90.0	WPE	91.5	BSW

* Sand-cast zinc anodes containing 0.3 to 0.5% Al and 0.25% Hg.
** Bath compositions given in Table No. 1.
† All deposits at least 400 ampere minutes = 0.0005 in. at 100% efficiency.
‡ Air agitation.

Legend: BSW—best of baths I to IV, smooth white
BTE—best of baths I to IV, treed edges
WS—white, smooth
DS—dark smooth
DTE—dark, treed edges
GS—gray, smooth
GTE—gray, treed edges
WTE—white, treed edges
WPT—white, pitted, treed
WPE—white, pitted edges

The anode efficiency in all four baths is practically 99 per cent and is independent of the current density within the above limits.

The anode polarization is practically unaffected by the current density. It is less than 0.1 volt for Baths I and II and 0.23 and 0.14 volt for Baths III and IV, respectively.

The alloy anode produces no sludge in any of the baths under these conditions of operation.

The cathode efficiency for any one bath varies somewhat irregularly with changes in current density. The

TABLE NO. VII.—Variation of Bath Voltage and Acidity with Current Density and Bath Composition in Acid Zinc Baths Using Al-Hg-Zn Anodes*

Bath Compositions ** Baths I-IV		I		II		III		IV	
		Electrometric pH at Start † 3.34		3.34		3.34		3.34	
Temp., Deg. C.	C.D. Anode, amp. per sq. ft.	Bath, volts	pH change	Bath, volts	pH change	Bath, volts	pH change	Bath, volts	pH change
50 ± 2	100	3.40	0.50	3.50	1.78	2.70	...	2.90	0.09
50 ± 2	150	4.25	0.26	4.50	0.85	3.30	0.42	3.90	0.50
50 ± 2	200‡	5.10	—0.47	5.30	—1.8	4.20	—1.6	4.75	0.43
50 ± 2	300	8.00	0.04	...	...	6.6	0.03	...	...

* Sand-cast anodes containing not over 0.5% Al and 0.25% Hg.
** Bath compositions in Table No. I. Bath volume = 3 and 4 L in cold and hot baths respectively.
† Colorimetric pH 4.0 = (quin hydrone) pH 3.34 at room temperature. All pH values positive unless otherwise shown.
‡ 0.2 cc/L concentrated sulphuric acid added during run.

average efficiency for all three current values shows that Baths II and III (Table VIII) have about the same cathode efficiency within the range of current density covered. The efficiency of Bath I is slightly less and of Bath IV slightly higher. In comparing the anode and cathode efficiencies in the same bath (Tables V and VI) at current densities of 100 and 150 amp. per sq. ft. (10.8 and 16.2 amp. per sq. dm.), it will be

noted that the cathode efficiencies are somewhat lower than the anode efficiency, which is constant at about 99 per cent. It was thought that agitation of the solutions might bring these values closer together. In Table VI at 150 amp. per sq. ft. (16.2 amp. per sq. dm.), it is seen that air agitation raised the cathode

TABLE NO. VIII.—Comparison of Baths I to IV*

Bath No.	I	II	III	IV
Anode Efficiency,** %	99.0	98.7	99.0	99.0
Anode Polarization, volts	0.09	0.08	0.23	0.14
Cathode Efficiency,† %	89.7	92.8	92.6	93.6
Character of Deposit	Best white	Dark	Gray	Good white
Bath Voltage, volts	3.4-5.1	3.5-5.3	2.7-4.2	2.9-4.75

* Average of values at four current densities, at 50 deg. C. (122 deg. F.).
pH = 4.0 colorimetric or 3.34 quinhydrone at room temperature.
** 100, 150, 200, and 300 amp. per sq. ft. (10.8, 16.2, 21.6, and 32.4 amp. per sq. dm.).
† 80, 100, 150 amp. per sq. ft. (8.6, 10.8 and 16.2 amp. per sq. dm.).

efficiency in Bath I, but lowered it in the other three baths.

In Table VIII it is seen that deposits from Bath I are best as to color and structure. Bath IV is next best, while Baths II and III give dark and gray deposits, respectively.

The bath voltages vary between the limits shown in Table VIII, the highest voltage in each bath naturally being required at the highest current density.

Due to inaccuracies in maintaining a constant pH by adjustment during the short time runs at these high current densities, the changes in pH recorded in Table VII are too variable to permit any valid comparison to be made.

Conclusions

The Al-Hg-Zn anode is capable of operation at current densities at least as high as 300 amp. per sq. ft. (32.4 amp. per sq. dm.), with negligible polarization and with no sludging, in commercial acid baths operated at a pH of 4.0 and at a temperature of 50 deg. C. (122 deg. F.). The anode efficiency is 99 per cent, being practically independent of current density, temperature and of the bath compositions studied.

The cathode current density cannot greatly exceed 100 amp. per sq. ft. (10.6 amp. per sq. dm.) in still plating if satisfactory deposits are to be obtained with the baths and conditions of operation used in this investigation. The cathode efficiency under these conditions will be somewhat less than the anode efficiency of 99 per cent. The difference may possibly be altered by the use of agitation but, where dragout is appreciable, the above difference may be an advantage. Rapid motion of the work being plated, such as rapidly drawing wire through a bath, may be expected to yield satisfactory deposits at still higher cathode current densities. Furthermore, a wire is simpler to plate since it does not present the problem of ununiform current distribution that rectangular sheets offer. The so-called "edge effect" is therefore absent on wire. The bath compositions giving the best deposits at the high current densities employed were I and IV.

Comparison of Al-Hg-Zn and Zinc Anodes

In order to fairly judge the results obtained for the alloy anode, one should know what a straight zinc anode of commercial purity would do under similar conditions. A comparison has therefore been made with straight zinc in Table IX at current densities of 100 and 200 amp. per sq. ft. (10.8 and 21.6 amp. per sq. dm.). At the former current density the straight zinc anode has an efficiency over 100 per cent in all four baths, indicating chemical solution or sludging of the anode. The highest value is 101.3 per cent in the Na_2SO_4 Bath No. III. At the latter current density the anode efficiency for zinc is somewhat lower, varying from 100.9 per cent in Bath I to 94.0 per cent in Bath IV. Under the same conditions the Al-Hg-Zn anode has the constant efficiency of about 99 per cent.

The efficiency for both anodes is also given for Bath I at a current density of 20 amp. per sq. ft. (2.16 amp. per sq. dm.) and room temperature. Under these conditions the efficiency for the zinc anode (106.9 per cent) is much higher than the value for the alloy anode. It is also considerably higher than the value for zinc at the higher current densities.

TABLE IX.—Comparison of Al-Hg-Zn and Commercially Pure Zinc Anodes in $ZnSO_4$ Baths

Bath Composition		Temp. average		I		II		III		IV		Remarks	
Anode Composition	C.D., amp. per sq. ft.	Deg. C.	Deg. F.	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Eff., per cent	Polarization, volts	Sludging	Anode Film
Al-Hg-Zn	20†	25	77	99.3	0.02	...	...	...	...	...	...	None	None
Zinc	20†	25	77	106.9	0.02	...	...	...	...	...	...	Metallics	Black
Al-Hg-Zn	100	50	122	99.2	0.09	99.0	0.06	99.2	0.23	99.1	0.14	None	Bright Hg
Zinc	100	50	122	101.0	...	100.1	...	101.3	...	100.8	...	Metallics	Black
Al-Hg-Zn	200	50	122	99.2	0.07	98.9	0.10	99.1	0.22	99.1	0.14	None	Bright Hg
Zinc	200	50	122	100.9	...	99.0	...	100.6	...	94.0	...	Metallics	Black
Al-Hg-Zn	300	52	126	98.7	...	98.0	...	98.4	...	99.1	...	None	Bright Hg

* Containing not less than 99.93% zinc.
† Data taken from Hogaboom & Graham, Trans. Electrochem. Soc., 62, 54 (1932).

Of more significance than the efficiencies is the sludging tendency, since this invariably affects the value of the deposit. While the alloy anode produces no sludge, the straight zinc anode gives an abundance of fine floating metallic particles.

It is also interesting to note that the alloy anode has an adherent bright mercury film, especially at the higher current densities, which undoubtedly contributes to the uniformity of behavior in the several baths studied. The film on the straight zinc anode in all four baths is thin, black and easily removed.

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Drawing by Charles Perry Weimer

1936 MARCHES ON!

Large Single Crystals of Copper

Simplified Method for Their Preparation

By J. G. THOMPSON

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Publication Approved by the Director

SYNOPSIS: Single crystals of copper several inches in length and up to 2½ in. in diameter were prepared by a simplification of the Bridgman method of progressive solidification of the molten metal. The vertical Arsem furnace permits slow cooling of a melt from the bottom of the crucible upward without requiring the use of any moving parts.

The orientation of the single crystals of copper was determined by the use of etchants according to the method of Hausser and Scholz.

SEVERAL INGOTS of oxygen-free copper, which had been prepared in a laboratory vacuum furnace of the Arsem type, weer observed to be very coarsely crystalline. To prepare one of these ingots, 7 or 8 lb. of electrolytic copper was melted in a graphite shell in the furnace, the heating unit of which was a spiral resistor of graphite. The pressure of the atmosphere in the furnace was maintained at 2 to 3 mm of mercury. The copper was held in the molten state long enough to permit deoxidation by the reaction between copper oxide and the graphite shell. The power was then shut off and the contents of the crucible were allowed to solidify in the furnace.

Pickling of the ingot in dilute nitric acid served the dual purpose of cleaning the surface and of indicating the size of the crystals. Under these conditions, crystals 3 in. or more in length, were frequently observed in the lower portions of the cylindrical ingots. This suggested the possibility of producing single-crystal specimens by a simplification of the Bridgman method[1] of progressive solidification from a single nucleus in the constricted lower portion of a cylindrical container.

In the Bridgman method the container and the inclosed metal specimen are lowered through a furnace which is maintained at a temperature above the melting point of the metal. Solidification usually begins from a single nucleus in the constricted lower portion of the cylindrical crucible and the resulting crystal may continue to grow until it reaches the full length of the crucible. The rate of travel through the furnace must be controlled to minimize the possibility of the formation of additional nuclei.

Progressive Cooling in the Arsem Furnace

The construction of the vertical Arsem furnace is such that progressive cooling of a crucible, from the bottom upwards, can be attained without the use of moving parts. The spiral heating element is mounted at the lower end in a graphite block and electrical connections are made through water-cooled clamps. In this particular furnace the temperature of the resistor is controlled by regulating the power supplied to the furnace from a 25 kva step-down transformer operated on 2300 volts at 60 cycles. Manipulation of a number of switches connected to taps in the secondary winding of the transformer permits delivery of power to the furnace at any desired voltage in 1-volt steps between 1 and 61 volts. The graphite block and the lower end

Fig. 1. Graphite Crucible for the Preparation of Single-Crystal Rods.

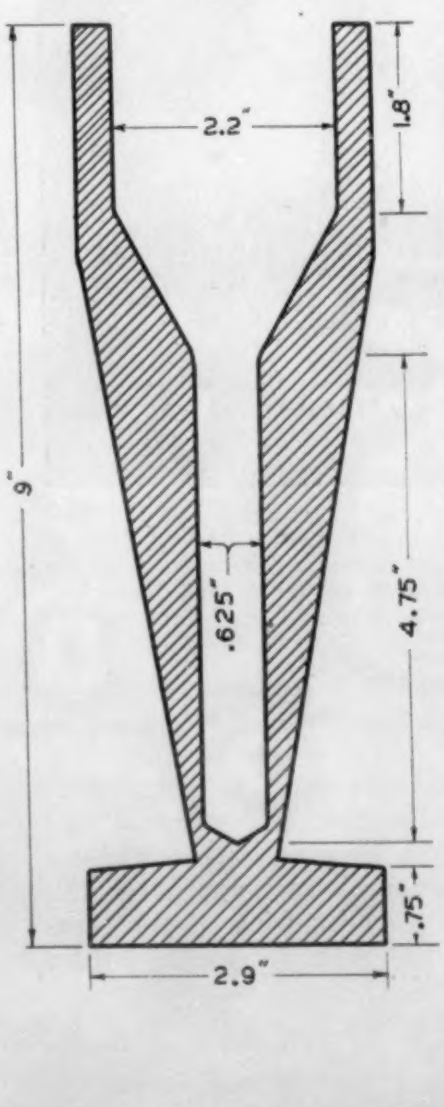
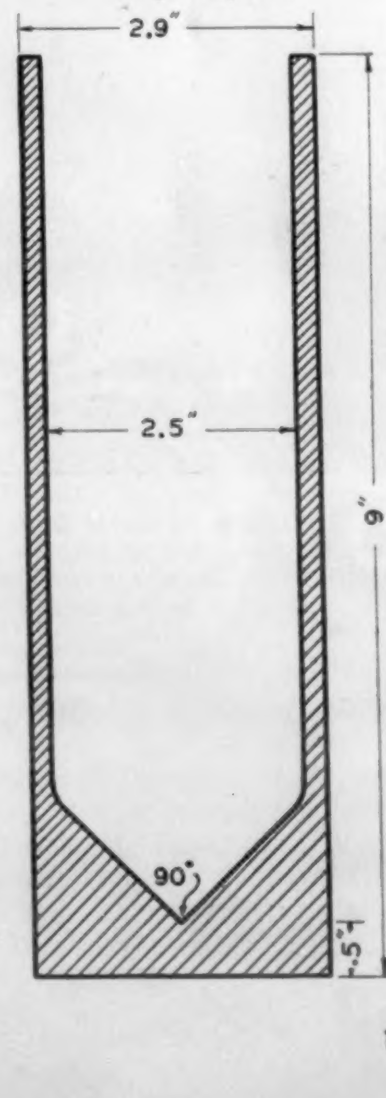


Fig 2. Graphite Crucible for the Preparation of Large Single Crystals.



of the spiral are always at a lower temperature than the central portion of the resistor, as a result of the circulation of water through the clamps. Ordinarily it is customary to support the crucible on an insulating block to avoid chilling the bottom of the crucible, but in the present work where the cooling effect was desired the crucible was mounted directly on the graphite block.

The preparation of rods 0.625 in. in diameter with a monocrystalline portion 2.5 to 3.5 in. in length was found to be relatively simple. The crucible which was used is shown in Fig. 1. Single crystals 2 or 3 in. long were ample for some cold rolling experiments[2] which were in progress at that time, consequently no particular attention was paid to the design of the crucible or to variations of operating conditions which might have resulted in the production of longer monocrystalline rods.

Graphite Crucible for Large Crystals

Subsequently, it was decided to attempt the preparation of single crystals of larger diameter and a graphite crucible was constructed with the dimensions shown in Fig. 2. This crucible, which was about as large as could be inserted within the resistor, has a capacity of approximately 3500 grams of molten copper. The crucible was covered to decrease thermal radiation from the surface of the melt. A hole in the cover permitted visual observations. The usual procedure was to charge the crucible with deoxidized copper, heat until the copper was thoroughly molten, lower the power input until the temperature of the hot zone of the resistor was slightly below the melting point of copper, and then maintain this temperature for 1.5 to 2 hrs., after which the power was cut off.

The preparation of each ingot required approximately 4 hr., about 2 hr. for the melting and an equal

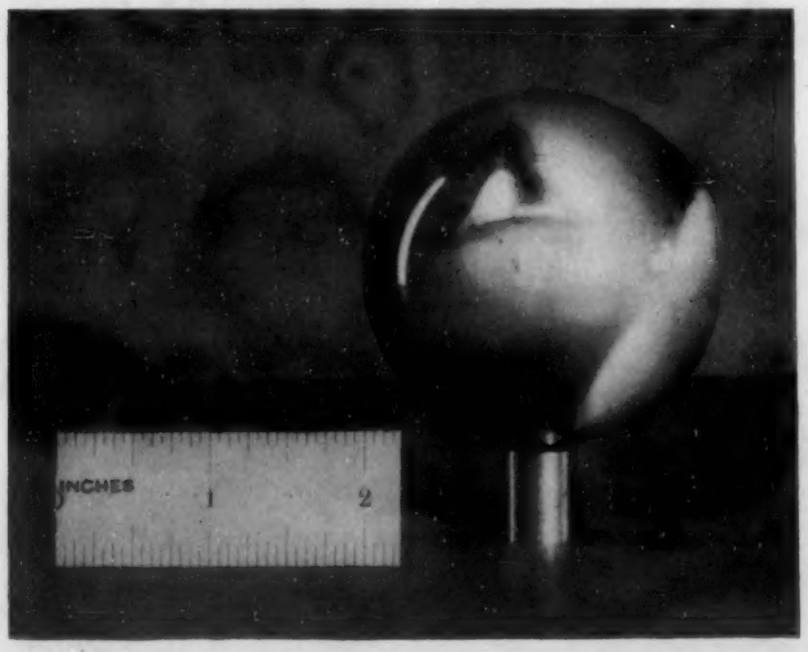


Fig. 3. Sphere 2¼ in. in Diameter, Cut from a Single Crystal of Copper. Note that etching to determine the orientation of the crystal lattice likewise reveals evidence of "linear" or "dendritic" structure in this single crystal of rather pure metal.

time for the cooling procedure with power still on. After the power was cut off, the crucible was allowed to remain in the furnace until the following morning. The temperature of the metal was estimated from previous experience with the voltage and time regulations for the furnace. Pyrometric observations were not attempted as the presence of a thermo-couple in the melt was not desired, and determinations of the temperature of the surface of the melt, by means of an optical pyrometer, would have been of doubtful value.

On the second attempt, an ingot was obtained which consisted of only two crystals. One crystal started in the conical bottom of the crucible but, by the time it had progressed approximately 2 in. up the vertical walls, another crystal started on one side of the crucible. The two crystals thereafter continued to grow, the first crystal decreasing in cross section and the second increasing as the top of the ingot was approached. The upper portions of the ingot, which contained an appreciable shrinkage cavity, consisted almost entirely of the second crystal. The single-crystal sphere, 2¼ in. in diameter, shown in Fig. 3, was machined from the lower portions of this ingot and was part of an exhibit of the National Bureau of Standards at the Century of Progress in Chicago during 1933 and 1934.

Revealing the Orientation of the Crystals

The orientation of the crystal is revealed by the etched pattern which is indicated in Fig. 3, but is shown to better advantage under the lighting arrangements used to obtain Fig. 4. In the etched pattern, resembling a four-leaved clover, each leaf shows the traces of a family of dodecahedral (110) planes on the spherical surface; the center of the clover is the center of a cube face. This method of determining the crystal orientation in a spherical specimen was developed by Hausser and Scholz[3] during an extended study of the action of various etchants on single crystals of copper, silver, lead, and some of their alloys. For copper, it was found that several etchants were available, each having a selective action on the atoms of one or more planes of the crystal. On silver crystals, a composite pattern showing the location of several different planes could be developed but suitable etchants for lead were not discovered.

The characteristic patterns developed by these etchants permit a rather definite determination of the orientation of the atomic lattice in the single-crystal specimen. Identification of a characteristic pattern and determination of the orientation are quite as exact on an etched hemispherical surface as on a complete sphere. This method of determining orientation is decidedly more convenient than X-ray examination, but the latter is preferable for precise determinations.

How the Etching Was Done

In the present experiments with copper, etching of the spherical specimens and of the hemispherical ends of rod specimens was done according to the procedure of Hausser and Scholz, as follows:

> The specimen was first given a "blank etch" treatment with a solution consisting of equal parts of distilled water and concentrated nitric acid, plus a few crystals of ferric chloride. This reagent is non-selective in its action; it dissolves the surface uniformly and therefore can be used to remove the results of other etching treatments or the work-hardened material which results from machining operations. After treatment with the blank etch, the dodecahedral (110) planes can be revealed by immersion in concentrated ammonium hydroxide to which hydrogen peroxide is added during the etching of the specimen; the octahedral (111) planes are revealed by etching with a hot 10 per cent aqueous solution of ammonium persulphate; and the cube (100) faces, by means of concentrated nitric acid which contains a little silver nitrate. In the etching to develop the cube faces some difficulty was encountered from staining but, on the whole, the development of any one of the three sets of planes could be satisfactorily accomplished. After the etched pattern had developed, the specimen was washed in running water and in alcohol, dried and coated with transparent lacquer.

After the production of the crystal from which the sphere shown in Fig. 3 was obtained, subsequent melts yielded similar results. In each case an additional crystal or crystals developed after the first crystal had grown approximately 2 in. up the vertical sides of the crucible. This indicated non-uniform cooling of the circumference of the crucible, which might have resulted either from a hot spot in the resistor or from improper centering of the crucible in the furnace. In an attempt to minimize and equalize lateral radiation the vertical sides of the crucible were lined with magnesia. When the crucible was fired the lining, which was about 0.25 in. thick, contracted and left a space of 1/8 to 1/16 in. between the lining and the graphite wall. This space was too narrow to be filled conveniently by packing with dry magnesia.

Approximately 2000 grams of copper was placed in the lined crucible and the cover was weighted to prevent the lining from floating on the molten copper. The time required both to melt the copper and subsequently to cool the melt was noticeably greater than in the previous experiments with an unlined crucible. Examination of the crucible after it had cooled showed that copper had penetrated under the lining and filled the space between the lining and the graphite shell to the same level as the top of the ingot within. The ingot weighed 1750 grams and consisted of a single crystal up to within less than 1 in. of the top, which was coarsely crystalline.

It is interesting to note that the copper shell, which surrounded the magnesia lining, likewise consisted mainly of a single crystal. Only the upper portions of the shell were polycrystalline, the upper limit of the single crystal in the shell corresponding closely with the top of the single crystal in the ingot. When the shell was split and flattened, the single-crystal portion of the sheet was roughly 6x2.5 in., varying in thickness from 1/16 to 1/8 in. Previous reports of the preparation of single-crystal hollow cylinders or sheets have not been noted. Apparently it should be possible with this method to produce single-crystal specimens ranging from solid cylinders to flat plates by varying the shape of the crucible.

Results of the Experiments

These experiments have demonstrated the ease with which large crystals may be produced by slow cooling from the molten state in a crucible arranged for cooling from the bottom upwards, conditions which are readily attained in a vertical Arsem furnace. The maximum length of single cystals obtainable by this method is determined by the distance from the bottom of the crucible to the level of the zone of maximum temperature in the graphite resistor. Ordinarily the zone of maximum temperature occurs at approximately the center of the resistor and, consequently, the maximum length of a single crystal is approximately one half of the length of the resistor.

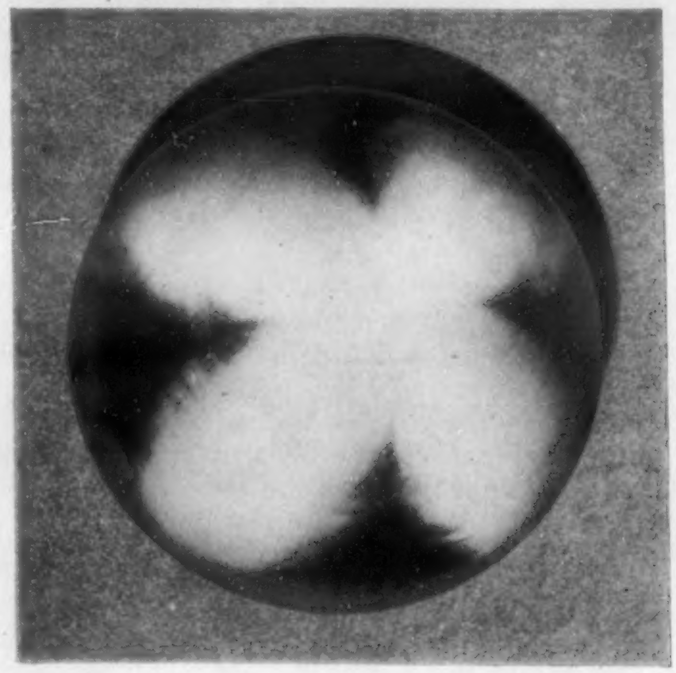


Fig. 4. Orientation of a Single Crystal of Copper May Be Determined by Selective Etching. This illustration shows the traces of four dodecahedral (110) planes on the spherical surface.

The experiments described in this paper were limited to copper but it is planned to extend the work to include other metals.

The author acknowledges his indebtedness to Louis Jorden for advice and to S. E. Sinclair and H. E. Francis for assistance in the experimental work.

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The Effect of Thallium on the Freezing Point of Lipowitz Alloy

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LIPOWITZ ALLOY very closely approximates in composition a quarternary eutectic alloy. It is composed of bismuth, lead, tin, and cadmium in the following respective ratios: 50:27:13:10. While various melting points ranging from 60 to 80 deg. C. have been reported for this alloy, the more recent investigations[1, 2, 3, 4] have fixed the melting point of the alloy at very close to 72 deg. C., and the freezing point at very close to 70 deg. C. The freezing point of this alloy is the lowest freezing point obtainable with any combination of the four metals involved since the alloy approximates the eutectic composition. If, however, the molten alloy is permitted to cool without stirring, undercooling results and the alloy may freeze as low as 60 to 65 deg. C.

To produce an alloy solidifying below 70 deg. C., mercury is sometimes added to Lipowitz Alloy. However, the addition of appreciable percentages of mercury produces an alloy which becomes pasty considerably above the freezing point, does not freeze sharply, loses much of its metallic luster, and property, and forms a brittle solid.

As has recently been shown by the senior writer[5], indium (M.P. 155 deg. C.) when added to Lipowitz Alloy lowers the freezing point of the latter markedly. Such an alloy, containing 18 per cent indium, freezes sharply at 46.7 deg. C. and melts sharply at 49.6 deg. C. Unlike the mercury alloy, the indium alloy retains its metallic properties. The alloy has properties similar to those of Lipowitz Alloy but does not oxidize as readily.

Thallium being a fusible metal (M.P. 303.5 deg. C.) might also be expected to lower the freezing point of Lipowitz Alloy, though the lowering should, of course, be much less than that produced by indium. Experiments were therefore conducted to determine the composition of the thallium-Lipowitz-alloy having the lowest freezing point.

A sample of Lipowitz Alloy was prepared from the purest C.P. metals obtainable. The melting and freezing points of the alloy were established, by means of melting and freezing curves, at 71.7 deg. C. and 69.7 deg. C. Fifty grams of the alloy were then placed in a pyrex test tube. The alloy was melted and an accurately weighed sample of thallium added. The tube and contents were heated above the melting point of thallium to insure the melting of the thallium. It was noted in the course of the work that thallium did not dissolve in molten Lipowitz Alloy at temperatures much below 160 deg. C. Indium, on the other hand, dissolves readily in molten Lipowitz Alloy at temperatures barely above the melting point of the Lipowitz Alloy. During the cooling process, the alloy was constantly stirred to insure complete mixing.

The tube containing the alloy was placed in a water bath of 800 cc. capacity which was in turn surrounded by an air bath. The water was stirred by means of a motor driver stirrer. The bath was permitted to cool by contact with the air of the room. Temperatures were recorded at one minute intervals by means of a thermometer, the bulb of which was immersed in the alloy. The thermometer was used to stir the

Fig. 1. Cooling Curves for Thallium Lipowitz Alloys.

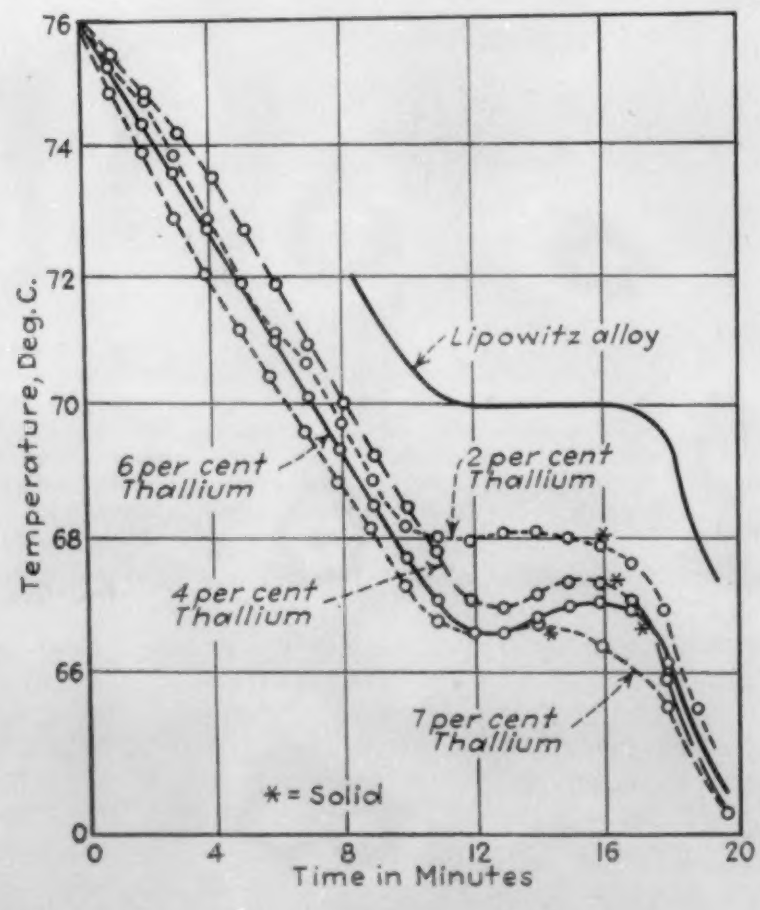
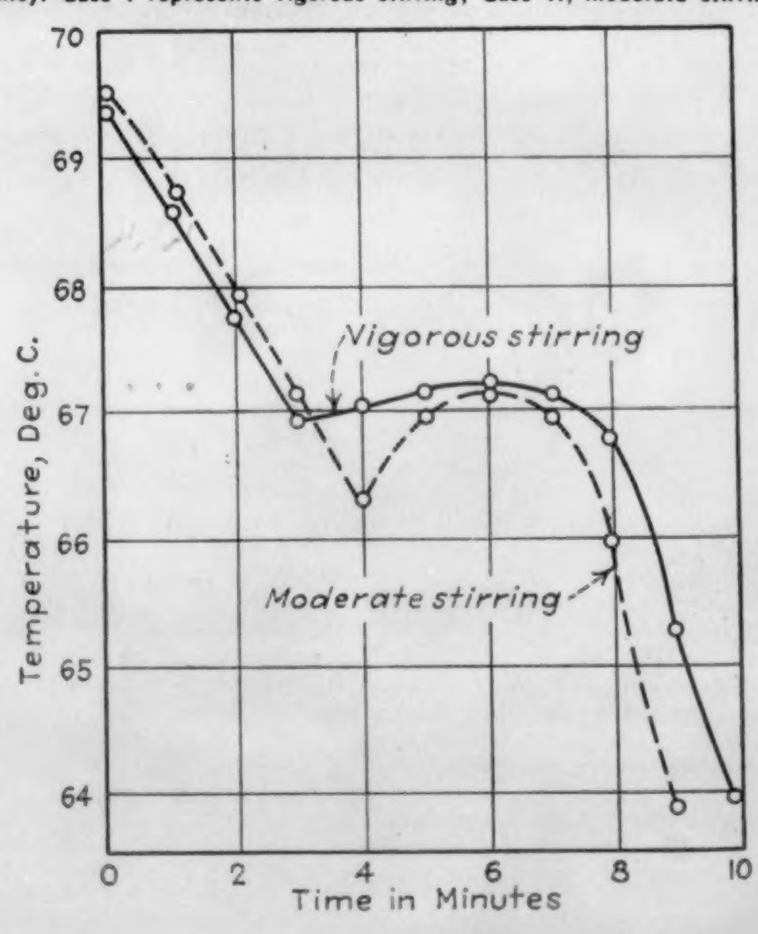


Fig. 2. Effect of Stirring on the Cooling Curves of Thallium Lipowitz Alloy. Case I represents vigorous stirring; Case II, moderate stirring.



molten alloy. The average deviation of this thermometer from a Bureau of Standards certified thermometer was + 0.2 deg. C. The deviation at the freezing-melting range of the alloys was + 0.2 deg. C.

Cooling curves were taken for alloys containing 1, 2, 3, 3.5, 4, 4.5, 5, 6, and 7 per cent of thallium. The percentages given are those obtained from the amounts of thallium added and not those of the final composition. No analyses were made of the solid alloys. Fig. 1 shows curves for alloys containing 2, 4, 6, and 7 per cent thallium. The curves indicate that the maximum lowering of the freezing point is obtained with an alloy containing 6 per cent thallium. Alloys containing 5 per cent thallium gave curves almost identical with those containing 6 per cent. Alloys containing less than 5 per cent thallium showed irregularities in the curve, indicating the solidifying of some ingredients before the minimum freezing point was reached. While the alloy containing 7 per cent thallium had the same solidifying temperature as that containing 6 per cent, it became slightly granular before the freezing point was reached and gave a somewhat irregular curve with a shortened horizontal portion. The lowest solidifying temperature obtained, that of the 6 per cent alloy, was 66.6 deg. C. This temperature is 3.1 deg. C. below the freezing point of the Lipowitz Alloy used.

While no cooling curve studies were made of alloys containing more than 7 per cent thallium, the approximate freezing and melting ranges of two alloys richer in thallium were determined with results as shown in the following table:

Percent Thallium	*Freezing Range Deg. C.*	*Melting Range Deg. C.*
10.50	81 to 67	66.5 to 68
18.10	115 to 66.7	66.7 to 68

In both of these alloys granular solids appeared far above the final solidifying point. However, the final freezing point is practically the same as that of the 6 per cent alloy.

Melting curves were determined for a number of the alloys. In each case, the horizontal portion of the curve lay about 2 deg. C. above the horizontal portion of the cooling curve for the same alloy.

One peculiarity of the thallium alloys, not noted to such a degree in the study of other fusible alloys, was the strong tendency of the alloys to undercool in spite of stirring. Fig. 2 shows two cooling curves for the same alloy containing 5 per cent thallium. Case I was stirred vigorously as it approached the freezing temperature while Case II was stirred moderately. No study was made of undercooling in the absence of stirring.

Because of the tendency of both cadmium and thallium to oxidize at elevated temperatures, care must be taken, in heating the alloy when dissolving the thallium, not to bring the temperature far above the melting point of thallium. Otherwise, brown oxides, presumably those of thallium and cadmium, are formed. With proper precautions little or no evident oxidation takes place.

The thallium alloy resembles Lipowitz Alloy in luster, ductility, malleability, and tendency to oxidize. The specific gravity, electrical properties and coefficient of expansion of the thallium alloys were not determined. Because of the relatively high price of thallium, the cost of the 6 per cent thallium alloy would be more than twice the cost of Lipowitz Alloy. While the addition of less than 2 per cent of indium produces as great a lowering of the freezing point of Lipowitz Alloy as does 6 per cent of thallium, yet because of the higher price of indium, the thallium alloy is by far the cheaper. Mercury alloys are less expensive than either but because of the many changes produced in the alloy by the addition of mercury, the product is often unsatisfactory.

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LETTER TO THE EDITOR

Shatter Cracks and Flakes in Steel

To the Editor: The writer has followed the abstracts of S. Epstein on "Recent Progress in Steel Making Reported from Germany" with great interest and has obtained much valuable information from this work. METALS & ALLOYS and Mr. Epstein both deserve our thanks for this work. There are three minor points on which some comment might be made.

First, in the August issue, page 221, Mr. Epstein says that "shatter cracks were revealed by Waring and Hofammann as early as 1919." The writer would also call attention to an article by M. H. Wickhorst (now deceased) on "Internal Fissures in New Rail," *Proceedings,* American Railway Engineering Association, Vol. 16, page 389, 1915, in which he describes and illustrates an investigation of 80 sections of new rails in which 13 internal cracks were located. The work was apparently done in the laboratories of the Illinois Steel Co. at Gary, Ind., as reported on page 349 of the preceding article, also by Mr. Wickhorst. At this early date the term shatter crack was not in general use so in the title of the paper these cracks were called "internal fissures."

Second, on page 222 of the same issue, Mr. Epstein states "To complete the demonstration (that hydrogen causes flakes) it would seem desirable to show by hydrogen analyses that commercial steels of a certain composition susceptible to flake, contain more hydrogen than steels of the same composition not susceptible to flake." This experiment was tried by the writer using the analytical method of Vacher and Jordan, on small samples machined from cracked and uncracked steel about two years after the steel was made. Under these conditions no appreciable amount of hydrogen was found in either type of steel. It must be remembered, however, that it is not hydrogen left in the steel which is being blamed for the cracks but rather the action of hydrogen in trying to escape. Also hydrogen will diffuse slowly through steel at room temperatures so it would seem that, if the above mentioned demonstration is to be successful, provision must be made to analyze the steel very soon after rolling or some method of chilling devised to insure that the steel retains its hydrogen until the sample is in the furnace for analysis.

Third, a logical explanation of why hydrogen should cause the cracks to form at 200 deg. C. or 400 deg. F. is found in the article "The Embrittlement of Boiler Steel II" by Everett P. Partridge and W. C. Schroeder in the July, 1935, issue of METALS & ALLOYS, page 187, in which they report: "All of these investigators noted that the rate at which hydrogen diffused through steel increased with increase in temperature. In this connection it may be noted that the velocity of diffusion of hydrogen through iron has been found to increase suddenly at about 390 deg. F. (200 deg. C.)." If this phenomena works in the opposite direction it seems quite logical that the highest pressure produced by the gas would develop at this temperature during cooling and this would be the temperature range at which cracking occurs. The writer is familiar with two other separate investigations which have established this range as the temperature at which internal cracks develop in certain steels.

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BERYLLIUM-IRON ALLOYS

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Fig. 1. (Left) Beryllium Steel about 1 Per Cent Be and 4.77 Per Cent Ni. Fine structure.

Fig. 2. (Right) About 1 Per Cent Be with no Ni. Coarse structure.

THE BINARY SERIES were first investigated by Oesterheld, who established the equilibrium diagram[1]. The age-hardening of these alloys was disclosed by Masing and Dahl[2]. The effect of nickel in such alloys was described by Kroll[3]. A good contribution to the theory of the age-hardening of quaternary Fe-Be alloys is given by Bennek and Schafmeister[4]. The influence of the Be on the gamma field in the binary alloys is described by Wever[5]. Another publication refers to the effect of Be in carbon-bearing alloy steels[6]. Desulphurization of steel baths with Be is given by Kroll[7].

Development of Be-Fe Alloys

The binary alloys, as described by Masing and Dahl, are of a coarse and brittle structure. They are hardenable by precipitation hardening with 4 per cent Be up to a maximum of 460 Brinell. For technical purposes such alloys, closely resembling phosphorus and aluminum-iron, are valueless. By adding nickel, the coarse structure gradually disappears. An alloy with 4.77 per cent Ni and 1 per cent Be, has a fine grain similar to a good nickel-steel (Fig. 1), while the 1 per cent Be alloy without nickel is very coarse (Fig. 2). The addition of Ni has another effect, namely, that it reduces the amount of Be needed for obtaining the maximum hardness, this being much higher than that obtainable in the binary series, e.g., with 4.77 per cent Ni and 1 per cent Be, one secures a maximum hardness above 600 Brinell, combined with a smooth structure. (Fig. 1). The hardness, after quenching, ranges between 225 and 425 Brinell, so that these alloys can, to a certain extent, be machined in the quenched state. The action of Ni remains the same up to 21 per cent if 1 per cent Be is present. If more nickel is added, the hardness in the quenched and aged state falls suddenly to about 200 Brinell.

The data in the diagram (Fig. 3) refer to a quenching temperature of 1100 deg. C. and an aging at 450 deg. C. If quenching proceeds at a temperature of 1200 deg. C., and aging at 700 deg. C., the hardness values of the aged and quenched alloys up to a Ni content of 21 per cent are not essentially changed. With more than 21 per cent Ni, however, the hardness of the aged alloy increases to 350 Brinell. It may be observed that the irregularities in the alloys containing between 0 and 21 per cent Ni, must be attributed to a certain aging occurring during quenching, as the rapidity of chilling has an important influence on precipitation.

An explanation for the described phenomena, namely, that above 21 per cent Ni at a Be content of 1 per cent the hardness of the aged alloy is not influenced when quenching takes place at 1100 deg. C., and aging at 450 deg. C., while it rises considerably when aging and quenching are carried at higher temperatures, can be furnished, based on the observations of Bennek and Schafmeister[4] on quaternary Be steels. It appears that in the given concentrations up to 21 per cent Ni, alpha iron prevails, while the gamma state is predominant at higher Ni contents. There may be coexistence of the alpha and gamma iron state in the limit-concentrations. Beryllium must be much more soluble in the alpha than in the gamma iron. The hardening in the alpha state is also more powerful. Low temperatures are required for producing the hardening effect in the alpha irons, while the gamma irons need high quenching and aging temperatures.

In the alloys where the alpha and gamma state is coexistent, the age hardening is variable, depending on the temperature of the heat treatment, and two different hardness values may be obtained with the same alloy. The hardness due to the alpha state dis-

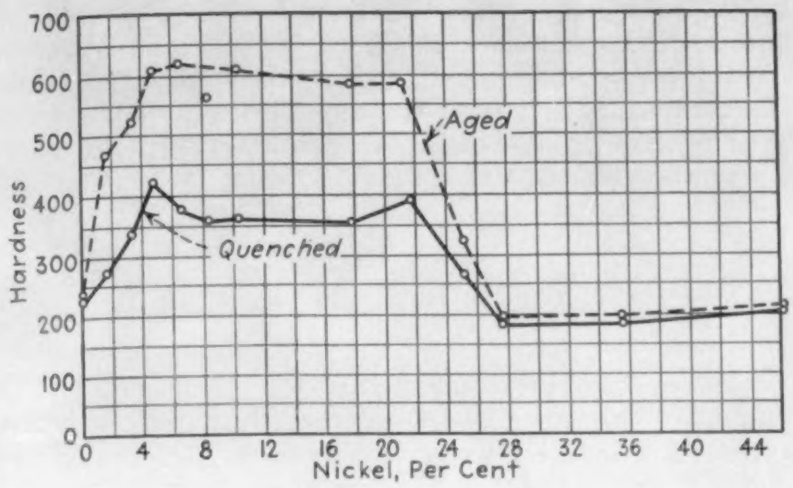


Fig. 3. Diagram for Heat Treatment of Ternary Fe-Be-Ni Alloys. Be, 1.02 Per Cent with Ni variable.

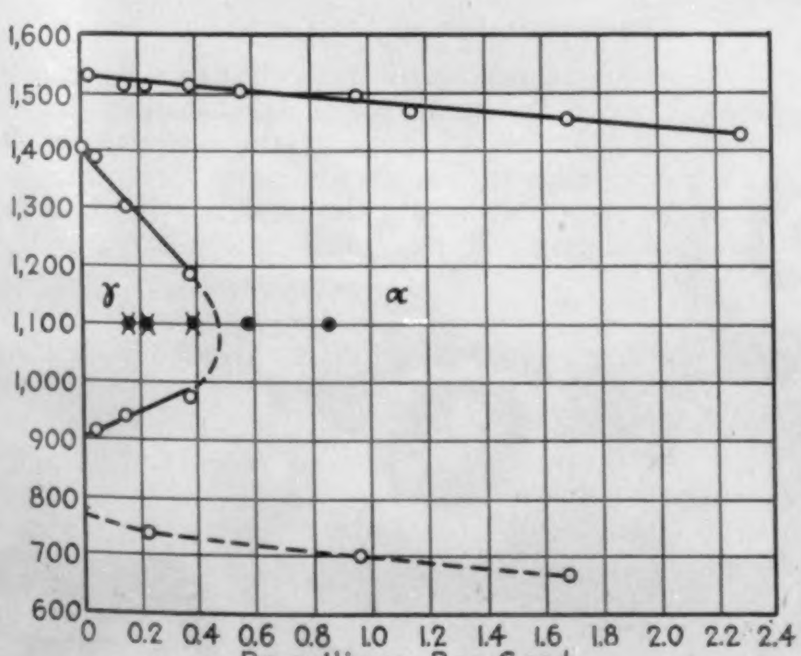
appears when aging at 700 deg. C., where the gamma state hardness appears. The pure Be gamma irons are not stable. After a prolonged heat treatment they can be completely decomposed, the austenitic structure disappearing and the alpha iron substituting for it. Wever[5] established that, when adding Be to iron, the gamma field is closed at about 0.5 per cent Be (Fig. 4). Nickel, on the contrary, increases the gamma field, so that Be and Ni are opposite in their action.

Effect of Adding Chromium

When adding chromium the tendency of producing alpha iron develops greatly, so that Cr and Be act in the same direction. There is no considerable increase in the obtainable Brinell hardness after aging by adding Cr, but a very good figure is obtained with an alloy containing 12 per cent Cr, 1.05 per cent Be, and 6.25 per cent Ni (Fig. 5). This alloy is hardenable up to 675 Brinell.

In these quaternary alloys, at a certain critical composition, the grain begins to be refined. This depends both on the Ni and on the Cr content, the first acting as a refining, the second as a coarsening agent. Beryllium steels, with 20 per cent Cr and 7 per cent Ni with 1 per cent Be, have to a certain extent a coarse structure, depending, however, on the amount of hot working. Without Be, these alloys would be austenitic, and they possess a considerable corrosion resistance. With Be, they are generally composed of

Fig. 4. Equilibrium Diagram of Fe-Be Alloys. (Wever)



alpha iron, hence of poorer chemical characteristics. They contain some new heterogenous constituents, and therefore their corrodibility must be considered. Nevertheless, their corrosion resistance compares favorably with that of ordinary C-Cr steels, as shown by Bennek and Schafmeister. They give values for this kind of alloy in comparison with a C-Cr steel in the Table.

Table of Comparison of Values of a C-Cr-Ni and a C-Cr-Ni-Be Steel

	Composition C	Cr	Ni	Be	Heat Treatment	Hardness
1)	0.34	13.7	0.45	...	Hardened	450
2)	0.11	19.0	8.7	1.19	950 deg. oil, 2 hrs., 500 deg.	505

Loss of weight in g/h m² in:

	HNO_3 (1.04) cold	hot	HNO_3 (1.23) cold	hot approx.	Acetic acid (1.06) cold	hot	HCl (1.017 cold)	H_2SO_4 (1.012) cold
1)	0.05	0.20	0.07	0.4	0.05	*	3 approx.	*
2)	0.012	0.24	...	4.5	0.024	0.59	0.16	40.5

* Not determined.

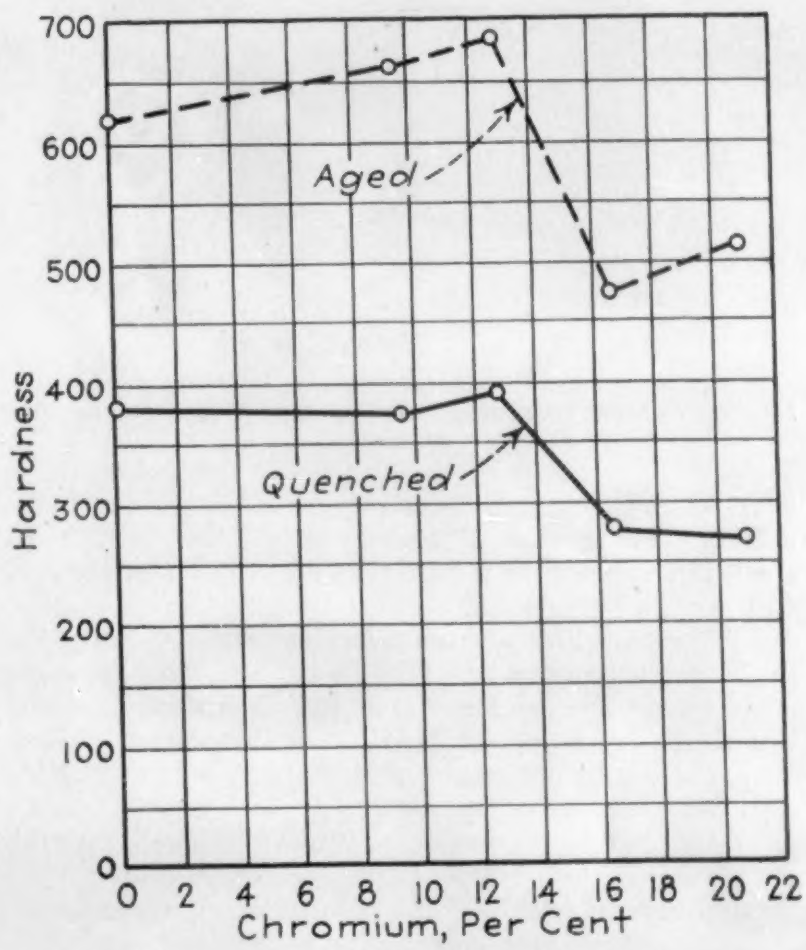


Fig. 5. Heat Treatment of Quaternary Fe-Ni-Be-Cr Alloys. Ni, 6.25 Per Cent; Be, 1.05 Per Cent; Cr, variable.

The corrosion resistance of rustless Be gamma steels is superior to that of the alpha steels, but, as described above, the maximum hardness only attains 350 Brinell. The results of Bennek and Schafmeister are not quite correct because they employed for their tests alloys with 0.1 per cent C and more. This element appears obnoxious in combination with Be in respect to corrosion[3]. It produces the readily decomposed compound Be_2C, the affinity of Be for C being most important. The great brittleness described by Bennek and by Dickenson and Hatfield[6] may be attributed largely to the presence of such a carbide. It can be observed that the Be alloys, so far disclosed, with higher C additions are readily attacked by reagents and show a pronounced hot shortness. Possibly a certain brittleness is inherent even in the completely carbon-free beryllium steels (that may be modified by adding other elements) but in general the applications of Be-steels are dependent on a raw material carefully controlled with regard to its C content. One of the best alloys, as actually revealed should contain about 12 per cent Cr, 7 per

Fig. 6a. A Steel Containing 1.20 Per Cent S and No Be. Hot shortness.

cent Ni, 1 per cent Be and no carbon. It has approximately the corrosion resistance of the best C-Cr steel, but can attain a hardness of 675 Brinell against 450 for the other steel.

In the beryllium-gamma iron steels the alloys of the Invar type may be of interest[3], e.g., an alloy with 36 per cent Ni and 1 per cent Be, with the remainder Fe, can be hardened up to 350 Brinell by aging. However, the coefficient of thermal expansion is somewhat unfavorably affected.

The magnetic properties produced in steels by adding Be are of two kinds: With greater amounts of Be in combination with Ni, one obtains magnet steels that perhaps are not yet sufficiently recognized, and it is possible that higher values for H_c and Br may be disclosed by suitable composition, heat treatment, and testing methods.

For permeability purposes, Be is a useful addition, as it improves the quality of ordinary permeability dynamo steels[8]. This may be attributed to a strong deoxidation, combined with the neutralization of C and N_2.

Affinity of Beryllium for S and O_2

One outstanding property of Be is its strong affinity for O_2 and S. It can be classed with the best deoxidizers ever used, and it displaces even Ca from its combination with S and O_2 at higher temperatures. This can be explained by the lower volatility of Be. Ca escapes as a gas at a temperature where Be remains stable[9]. The heat of formation and the volatility are given in the following table:

Heat of Formation		Vapor Pressure 760 mm. at deg. C.	
CaO	151.9	Ca	1439
BeO	145.3	Be	2970 [10]
MgO	143.9	Mg	1107
$1/3 Al_2O_3$	126.9	Al ... approx.	2400 [11]

From this it appears that the heat of formation of BeO and CaO are quite close together, but that Be volatilizes at a temperature twice as high as Ca. Therefore Ca can be reduced by Be up to an equilibrium depending on temperature and pressure. Due to these facts, Be is a much better cleanser for steel baths than Ca, since the latter is volatile at the operating temperature and does not dissolve in iron. As the affinity of Be for oxygen is very strong, one must be sure that no easily reducible oxides are present, either in the slag or on the walls of the furnace or ladle. However, in such cases, Be forms a skin of BeO that is highly refractory, and tends to prevent further attack. Under these conditions Al_2O_3 and MgO walls are quite sufficient for inhibiting the action of metallic Be. Liquid slags, however, containing SiO_2 or FeO, must be skimmed off before adding Be. The contact of the bath with the air does not produce heavy Be losses, as a thin skin of BeO on the metal surface, analogous to that observed in AL-Fe alloys, reduces the likelihood of severe oxidation. A suitable inert slag may be composed of fluorides with additions of alkaline earth metal oxides, such as a calcium fluoride-calcium oxide, or of glass.

Fig. 6c. The Same as Fig. 6b. Cold formed, no cracks.

Desulphurization of Steel with Beryllium

The desulphurization of steels by using Be is quite astonishing[7]. In small scale experiments, the insoluble BeS produced can remain occluded in the bath, e.g., similar to the compounds of Cr with S and Se in the free-cutting stainless steels. In this way alloys containing 0.76 per cent S in the shape of BeS

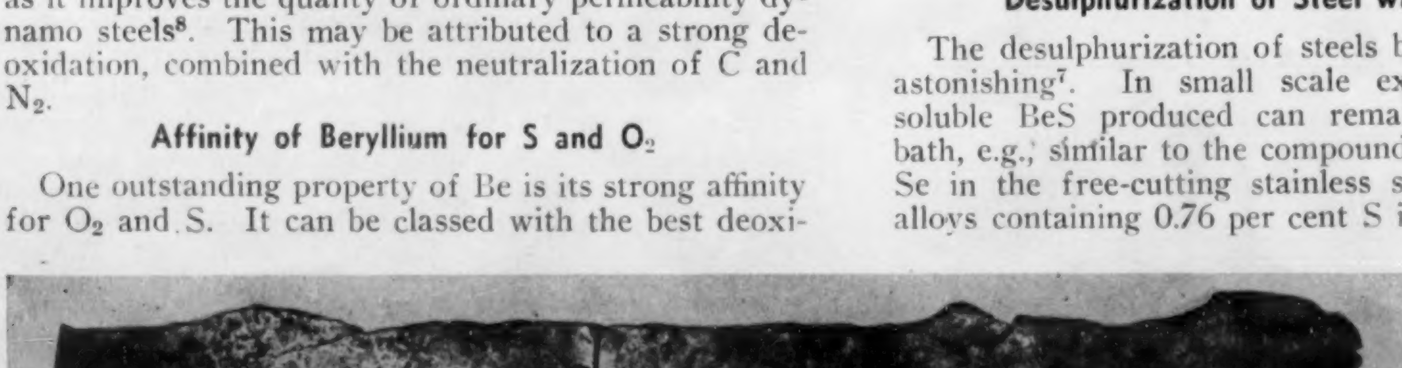


Fig. 6b. A Steel Containing 0.76 Per Cent S and 0.53 Per Cent Be. No hot shortness.

have been produced (Fig. 6 a, b, c). It is surprising that, in spite of such a high sulphur content, they could easily be rolled into thin sheets. Even ordinary pyrites, when desulphurized with Be, can be transformed into malleable iron. The affinity of Be for S resembles that of Cr, Zr, and Ti for the same element, but the action is much stronger. It is well known that Al is not a satisfactory desulphurizer of steel, insofar as hot shortness is concerned, so that the behavior of Be in this respect differs entirely from that of Al. As Be is soluble in iron, the desulphurizing action is more rapid and powerful than that of Ca or Li, which have been proposed for the same purpose. The percentages of Be needed for deoxidizing and desulphurizing steels are extremely small, due to its low atomic weight.

There remains much matter for investigation in beryllium steels. The first essential point that actually appears not to be sufficiently clear is the interreaction of Be with C, if the latter is present. The chemical nature of the carbide, Be_2C, is quite different from that of other stable carbides found in steels. It decomposes slowly in air, forming CH_4. Therefore if Be_2C is produced, one can expect a considerable diminution of the chemical resistance. Also the hot malleability is diminished. On the other hand, this compound may act as an insoluble impurity, eliminating carbon with the formation of "stable" carbides like Ti, Ta, and V. Small Be additions do not appear to act appreciably on C steels, but with more Be the hardness due to carbon disappears progressively.

The experiments performed by Dickenson and Hatfield[5] on carbon-bearing steels, employed in aircraft, after adding Be are, for the reasons described above, completely misleading. The action of Be on the hardening by carbon should be followed up in order to get a picture as to what extent Be_2C enters the solid solution without precipitation. The influence of Be on the hardening by carbon has not yet been investigated in respect to the depth of hardening, the coarseness of the grain, and the hardening sensibility. Beryllium may help penetration in surface hardening of steels in cyanide baths or in nitriding.

One new possibility is that of producing steels with a hard core and a very hard surface resulting from a combined age hardening, nitriding treatment in an atmosphere of ammonia. The action of Be on P, As, and N_2 in steels is not yet cleared up, and there is the possibility that Be forms compounds with these impurities, neutralizing the same, and producing "age-stable" steels. Analogous to Si, beryllium may be a useful element in Ni steels for increasing the elasticity of springs. This should be investigated together with the above-mentioned reasons for brittleness.

At the present time the price of Be has retarded its use in steels, commercial interest having centered more on alloys with copper and nickel. Investigation, however, should be based on a much lower Be price for the future, even if new uses for the time being may not actually be practical. Probably the most interesting immediate application will be in the deoxidation and desulphurization of special steels, with very low Be additions, under carefully controlled conditions.

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Inspection of Wire Rope

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IN PRACTICALLY EVERY CASE where it is used, wire rope has a heavy load to carry, and its sudden failure might easily prove serious. It is therefore wise to give all wire rope in service, periodical inspection. The frequency of such inspection will depend upon the severity and the importance of the service it is called upon to perform. Off-hand, one might say there are three outstanding factors to be checked against in making inspection of a wire rope. One is excess wear. The second is fatiguing and breaking of wires in the rope, while the third is security of all fastenings.

Excess wear may be caused by soft sheaves, by lack of proper lubrication, or by presence of abrasive in the air where the rope operates. The last two factors mentioned are interrelated, for what is proper lubrication in the place where much abrasive is present will not be proper where the air is relatively clean. Where the air is clean, the rope should be kept well lubricated on the surface. Where there is abrasive in the air, excess oil or grease will entrap much of the abrasive, form a matrix for it, and soon one will have a fair semblance of a valve grinding compound on the rope.

Obviously, such a condition is to be avoided, and the natural means of avoiding it is to lubricate far more lightly but more frequently, and to provide periodical cleaning for the rope. Spray cleaning with compressed air is a good method for forcing abrasive from between the strands of the rope, so that it may be wiped away.

Where wear of excessive nature is due to soft sheaves, the imprint of the lay of the rope can be seen to some extent in the bottom of the sheave. The rope first wears the sheave, and cutting begins, after which the sheave also wears the harder rope. Soft sheaves must be replaced with harder ones.

Tests for security of fastenings simply call for a judicious use of the wrench, and where any fastening is suspicioned, it may be tried out by lifting a greater load with it than will actually be required in service. Broken wires in a rope are easily located by rubbing over it with a handful of waste, as the broken wire ends will pull threads out of the waste, thus marking their own location.

Progress Report of the Joint Investigation of Fissures in Railroad Rails

COMMENT by S. EPSTEIN

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THE PROGRESS REPORT of the joint investigation of transverse fissures in rails by the rail committee of the Association of American Railroads and the Rail Manufacturers' Technical Committee, Bulletin No. 376 of the American Railway Engineering Association, which has been conducted under Prof. H. F. Moore's direction at the University of Illinois since 1931, is the first to be issued in printed form and contains 21 pages of text and an appended bibliography.

Prof. Moore is the dean of American testing engineers and, in this unusually authoritative report, he has settled many of the long-standing controversial points in the transverse fissure problem, largely by the straightforward use of mechanical testing. This clearing of the ground is of especial value at this time, as it helps answer the question whether the slowly cooled or otherwise thermally treated rails, which the steel mills have begun to produce recently, will prove a fully effective and practical solution of the transverse fissure problem.

The report definitely indicates that the so-called "shatter cracks" in newly rolled rails are the cause of transverse fissures. Not all rails which have shatter cracks develop fissures, but the typical transverse fissures, starting from nuclei inside the rail and gradually extending outwardly as progressive fatigue fractures, occur only in rails which already have "shatter cracks" when newly laid in the tracks. When no such shatter cracks are present, no true fissures form. Not enough tests were made of slowly cooled rails to prove that shatter cracks never occur in properly slowly cooled rails. But if this is true, as it apparently is, then the evidence in the report would indicate that slow cooling of rails after rooling (or other thermal treatments which eliminate shatter cracks) should prove a very practical solution of the internal fissure problem.

An extended abstract of the report has appeared in *Metal Progress* for November, pages 46 to 52, so that the report itself need not be summarized here. It may be stated, however, that the report is given over virtually entirely to the testing phases of the investigation and what metallographic and metallurgical work was done is briefly dismissed under the heading "Miscellaneous Lines of Investigation."

No outstanding features of metallographic structure were found to be associated with the presence of shatter cracks in rails, although a number of shatter cracks were observed to spread from segregation streaks in the rails. Attempts to develop internal fissures in rails having segregation streaks, but no shatter cracks, were unsuccessful. No correlation was found between the nitrogen and oxygen content in rail steel and shatter cracks. So far no relation has been found between steel making practice and the production of shatter cracked rails.

A metallurgist may perhaps be pardoned for wishing that more metallurgical work had been done and reported in this investigation, particularly as the present most promising solution of the transverse fissure problem is the metallurgical recourse of slow cooling. The metallurgical shortcomings of the report are perhaps best indicated by the fact that not a single chemical analysis is given. Similarly, in dealing with the tests of slowly cooled rails, nothing is said about the manner of slow cooling, or the temperature range through which the rails were slowly cooled (the secondary low ductile range or the blue brittle range, or both), nor is there any suggestion as to why slow cooling through any special temperature range should be effective in eliminating shatter cracks.

It is well known that the majority of the rails rolled in American mills, cooled directly on hot beds (not slowly cooled), are free from shatter cracks, and that only an occasional heat shows such cracks. A primary aim of any research on the shatter crack problem should, therefore, be the discovery of the cause and cure of shatter cracks in these occasional heats. Recent advances in the physical chemistry of steel making indicate that equal acumen and persistence on the metallurgical end, as was displayed on the testing end of this investigation, should clear up the whole matter and bring a solution that might perhaps be less expensive than slow cooling and be applicable to other steel products as well as rails. Prof. Moore is no doubt working along these lines and a report of his results will be eagerly awaited.

Protecting Sling Chains from Sharp Corners

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IT IS OF GREAT importance to inspect sling chains periodically, repairing them and annealing them at such times as may be necessary, but the most careful inspection and repair will not lessen the importance of care in operation. Many sling chains are abused in various ways by the operators of various types of lifting equipment. One outstanding form of abuse is the lifting of sharp-cornered loads of great weight without protecting the chain from the corners of the load. It is no wonder that under such a condition, individual links are bent, gouged and cracked.

Either wood or a soft piece of metal can be used to advantage between the links of a sling chain and the sharp corner on a load. Three-quarter-round made from some tough wood such as elm or hickory, may well be kept in stock in a moderate amount for this purpose, where considerable work having sharp 90-deg. corners is handled.

However, where the work varies much in form, and sharp edges and corners of different angles and contours are continually encountered, something in the nature of heavy sheet lead is to be preferred. Such material, interposed between chain and load, will both adapt itself to the contour handled, and permit the chain to embed itself to an extent which will go far in protecting it. This takes on added importance when we consider that protection for the sling chain is indirect protection for the workman, not to mention lowered maintenance and chain scrapping costs.